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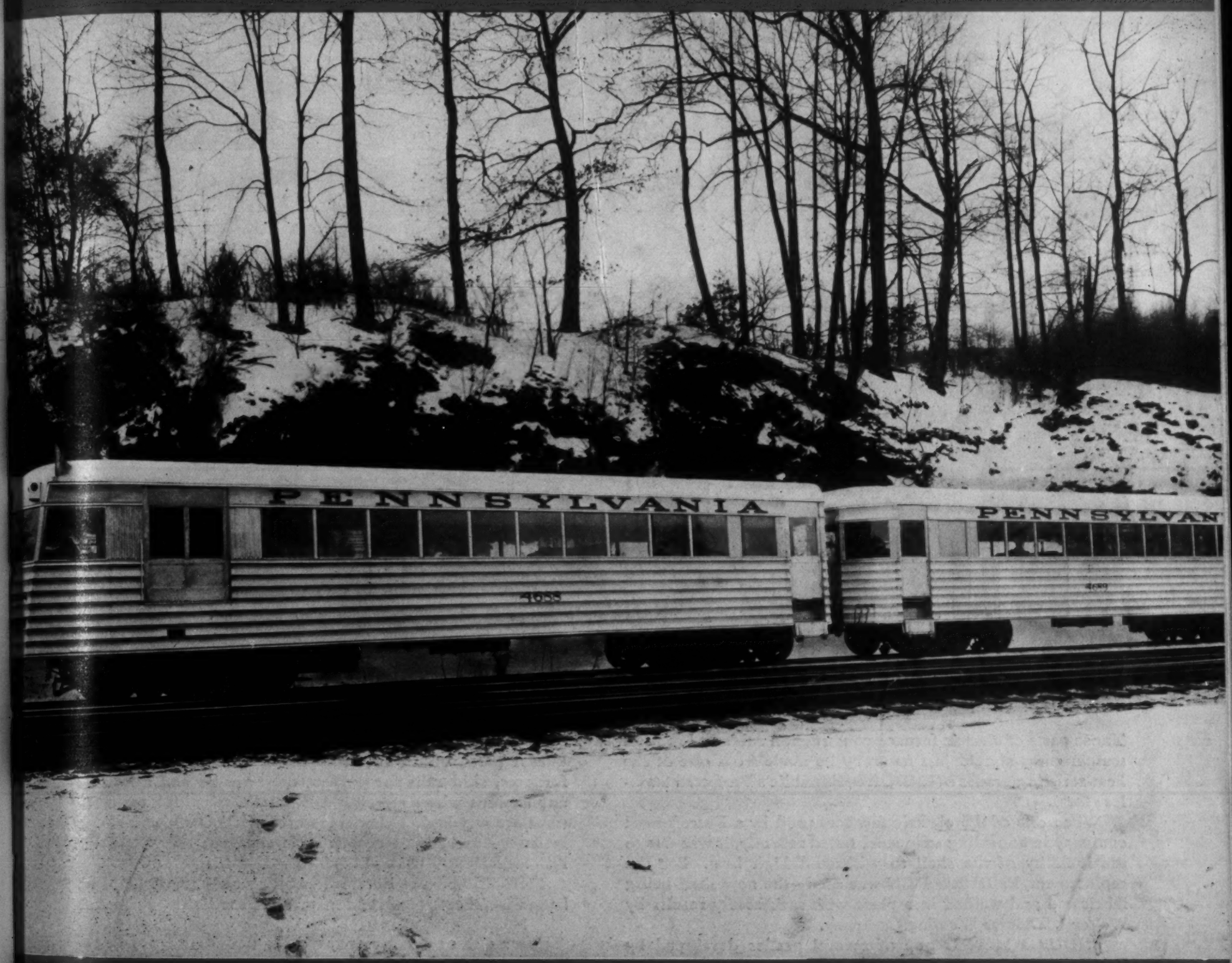
CURRENT METALLURGICAL ABSTRACTS



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VOLUME 4

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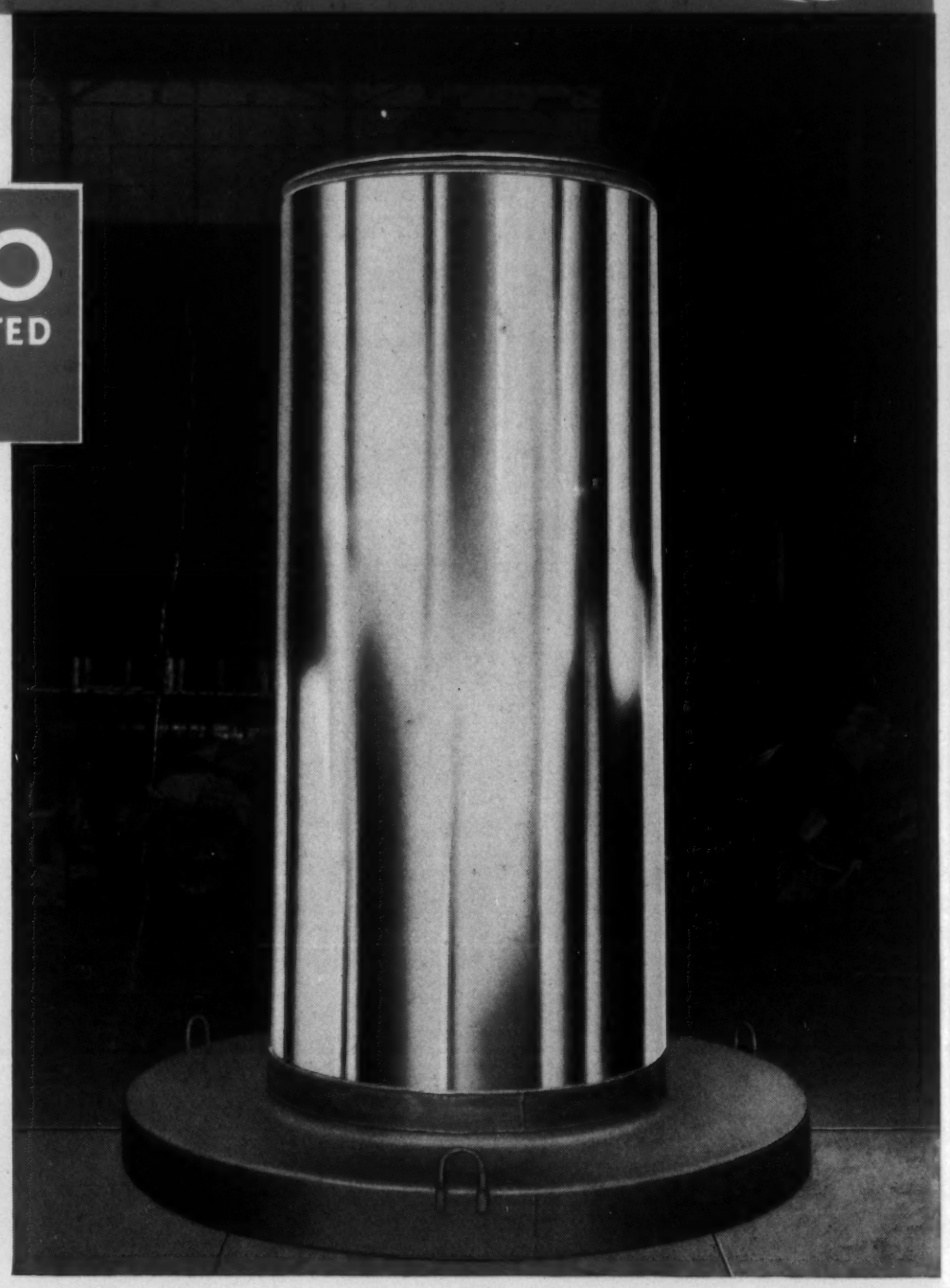

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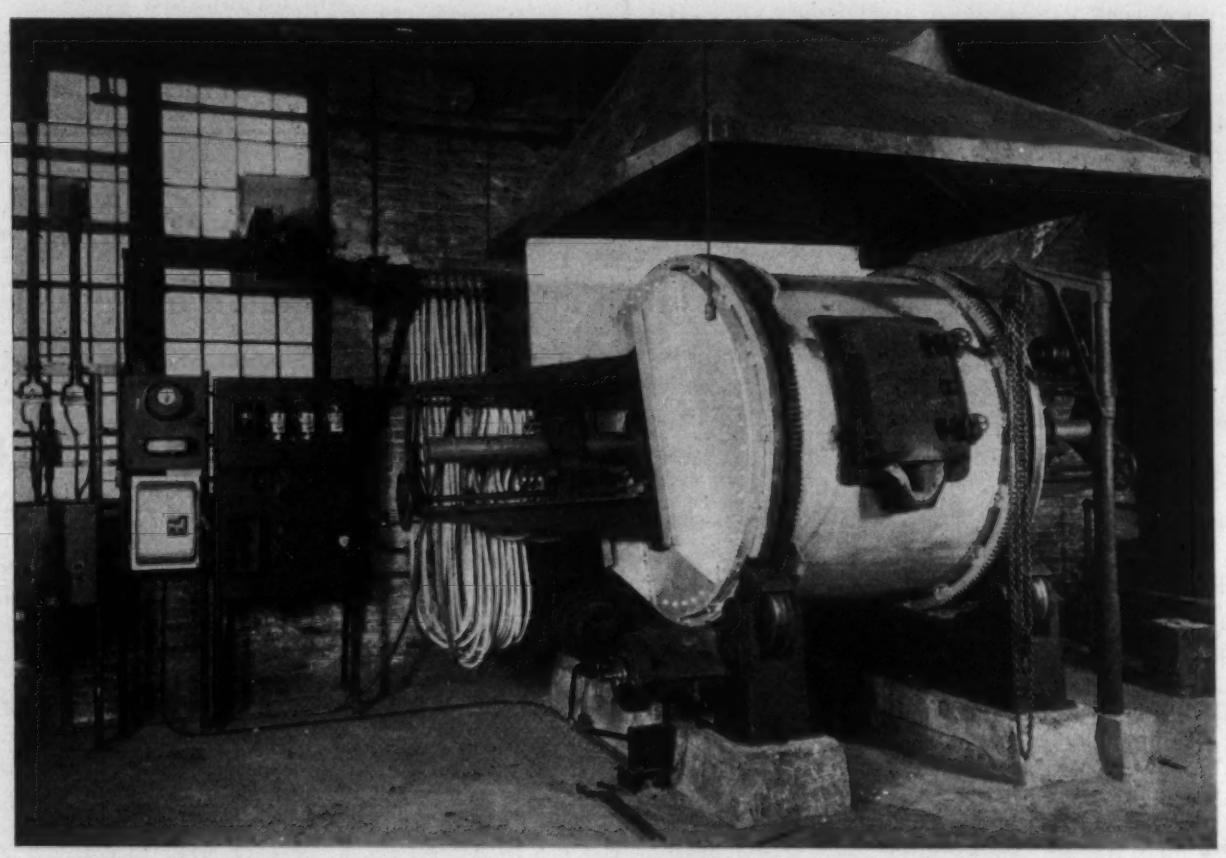

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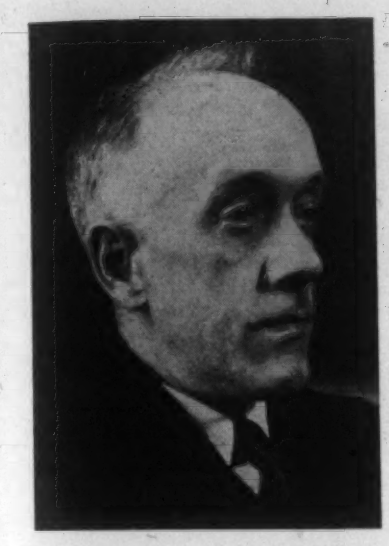
EDITORIAL COMMENT

Law Versus Gospel



WE DON'T often say or feel that "there ought to be a law" to remedy this or that, for there are too many laws now. But in the matter we have in mind, there has grown up what, in the absence of actual legal enactment, is the unofficial equivalent of a law, and, we think, a bad one.

In "Patent Law for Chemical and Metallurgical Industries," pages 163-164, Deller says that when the patentability of an invention is in doubt, a showing of commercial success and extensive use will throw the scales in favor of the patentee and influence the court in favoring the validity of the patent. This is particularly true if the patentee has succeeded in obtaining royalties from the public at large.

However, the citations given by writers on patent law for their justification of this doctrine are not to the Code, but to court decisions, i. e., not to laws created by legislative authority, but to judge-made gospel. Not all the decisions run in that direction, for Thropp's Sons Co. versus Sieberling (U. S. Reports, Vol. 264, October, 1923, pages 329-330) held that the mere fact that license fees were exacted, was no proof for the validity of a patent. Nevertheless, in the chromium plating case, recently carried up to the Supreme Court, the fact that licenses were taken out by many firms and an imposing total of license fees collected, seems, as we read the decisions, to have played a very large part in that case. The chromium case has been so much in the public eye that, unless there is positive legislative action to correct the situation, it will probably serve to strengthen what appears to be an unjustified and dangerous court point of view.

It is dangerous because it lends itself to a type of racketeering. Under the present weight of precedent, it is possible to set up a good defense for the validity of a patent merely by showing that a lot of people paid a lot of money to be licensed under, rather than to fight, it. That is, anything under which people can be persuaded to pay tribute must be prima facie a real technical advance, and one due to those who are exacting tribute.

This is absurd. If a patentee or someone who controls the patent can get some large firm to take a license for so small a fee as to make it cheaper than even to look into the validity of the patent, to say nothing of litigation, he can then go to other firms, asking larger fees, and because the A. B. C. Company, which is presumed to have good patent lawyers, is taking out a license, the D. E. F. and G. H. I. Companies may be persuaded to pay also. Thus, the snowball grows. Its size is no proof to any reasonable man that the patentee really invented the valuable part of the process.

(Continued on page 54)

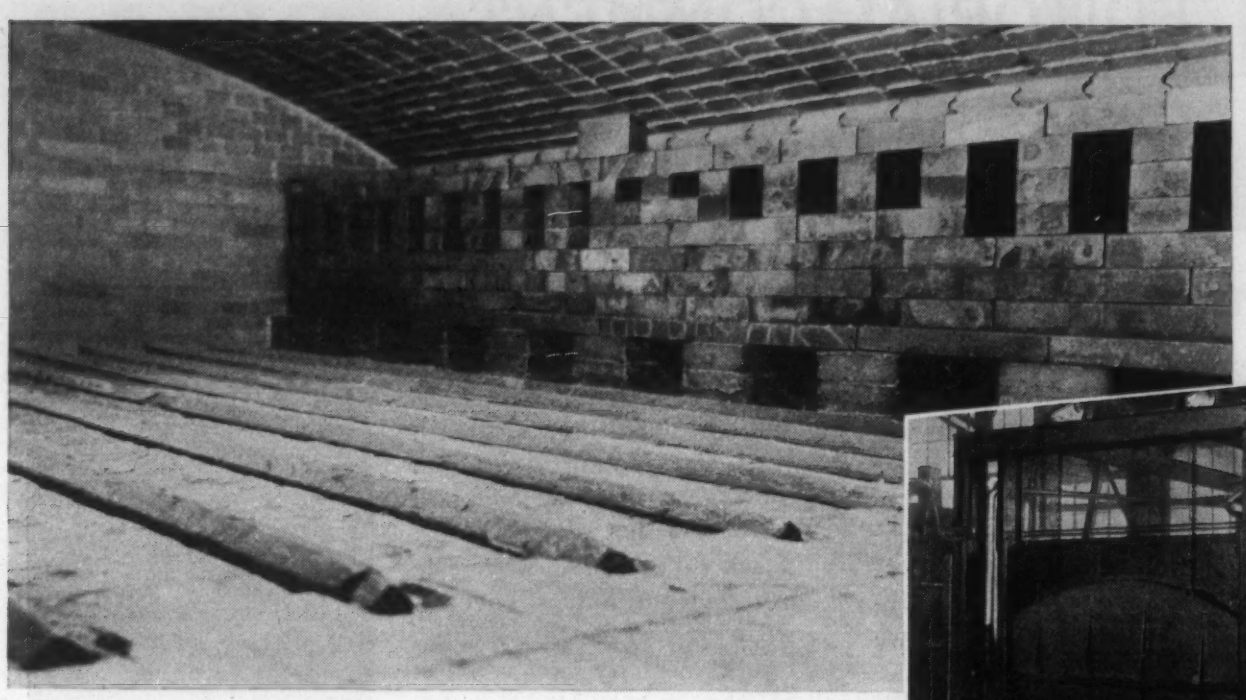
Surface Films

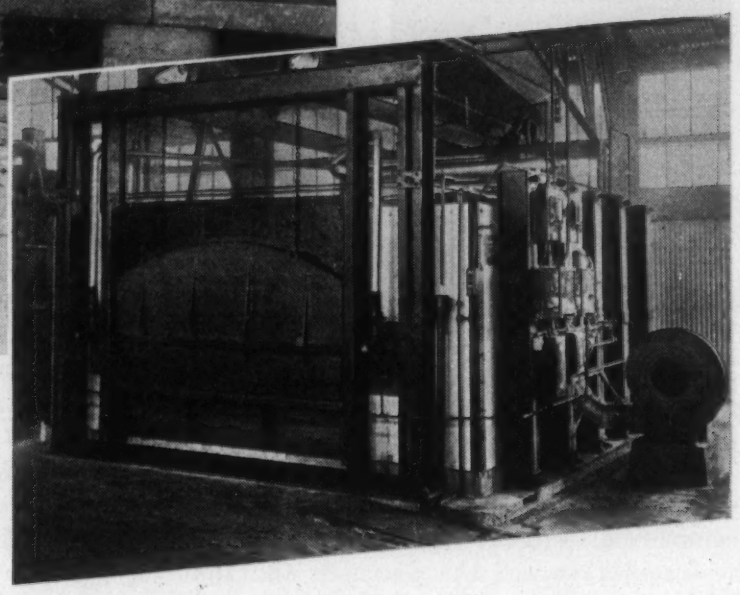
MUCH more is known about the internal properties and structure of metals than their surface condition, although it is now definitely known that the surface of common metals is not directly exposed to contact with the environment but is covered with a surface film. This film is usually some oxide of the metal. It has also been determined during the last few years that these films range from a few molecules up to 5×10^{-5} inches in thickness. The thin films may be invisible and are usually more durable than the thicker ones. On the other hand, rupture or lack of continuity in the primary film often determines the location of initial points of attack. The presence of such films has been proved by their effect on the electric potential of the metal, interference with the plating out of other elements from solution, and by actual removal of the film so that it becomes visible. In some cases the film has even been analyzed. Very little is known as to the bond between the metal and film, but it seems probable that the two are interconnected in a zone in which there is a gradual transition from the metal to the non-metallic film. However, it is expected that by application of more refined electron-diffraction methods, something definite will be learned as to the structure of the film and the nature of its bond to the base metal. Fundamental information of this kind is likely to be of considerable practical importance in improving protective coatings and in lengthening the life of metals.

Protective surface films are formed by reaction between the metal and its environment. In some cases, as in rust and heat resisting steels, a highly durable film forms automatically under many conditions of service. Under other conditions an oxidizing environment, such as strong nitric acid or sodium chromate, will lay down a passivating film on ordinary steel. In such cases however, the film is not permanently maintained except when in contact with the solution from which it is formed.

It has been frequently observed that the same metal will in some cases show remarkable durability (as illustrated in pictures of old wrought iron in the January 1933 issue of METALS & ALLOYS), and in other cases under apparently similar conditions will show comparatively short life. Some old wrought iron and steel pipe which apparently had not been protected with paint, left on the site of the Panama Canal by the French, was found by American engineers in a state of excellent preservation. The surface was covered with a dense skin somewhat similar to that found on copper steel exposed to the atmosphere. On removing this skin and exposing the pipe to the weather in the wet season, together with samples of modern pipe similarly treated, corrosion occurred in each case but no difference in the rate was

(Continued on page 48)





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Flakes in Nickel Chromium and Related Alloy Steels

By F. SAUERWALD,* H. GROSS and G. G. NEUENDORFF

A Correlated Abstract

F. Sauerwald

FLAKES in steel, as the literature on the subject proves, have caused extensive trouble in former years. They are still in evidence. This article endeavors to summarize our present knowledge and to give a brief account of our own investigations. The principal results of the most conclusive papers are tabulated in Table 1 according to the date of publication.† There will be found the observations of the individual authors and a brief statement of their views on the origin of such flakes.

H. Gross

G. G. Neuendorff

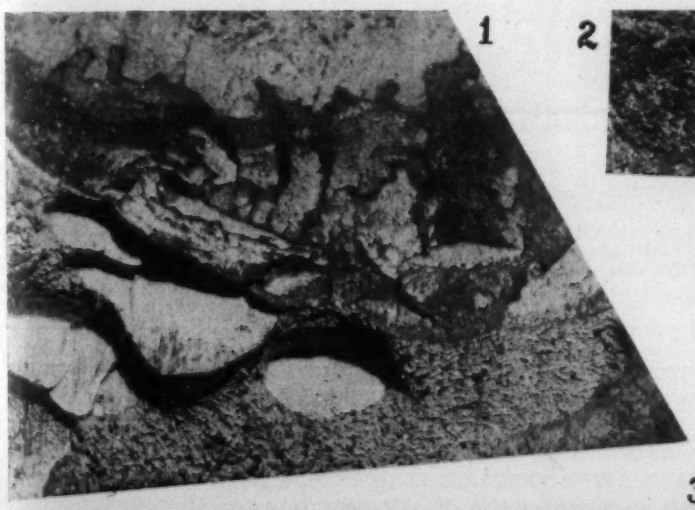
The following conclusions are to be drawn from the investigations cited. Flakes are the faces of cracks in steel. These cracks are apparent when the piece of steel is broken and the fracture follows these internal cracks. The appearance of such cracks differs from that of the remaining fracture, they are "flaky" in appearance.

Since the origin of the flakes, as will be seen later, is to be traced to a number of factors, they will not always be alike in appearance. It seems advisable to distinguish 2 principal types of flakes. The first type frequently possesses a finer grain (Figs. 1 & 2), and the second type a macroscopically coarser grain than the neighboring parts (Fig. 3). Neither nodular fracture nor woody fracture is identical with flakes, nor are the origins the same. Flakes are usually characterized by the following criteria:

1. They may be apparent in the ingot (see Fig. 2). The cracks which cause the flakes run parallel to the dendrites of the primary crystallites that froze last and are enriched in alloying elements (see Fig. 4). This appearance is most characteristic of flakes. In forging such ingots the cracks follow the

*Technische Hochschule, Breslau.

†There have not been included occurrences in high speed steel that seem to be of a different nature. See, for instance, Occurrence of fish-scale in high speed steel, Brophy, Harrington & Merrick, *Transactions American Society for Steel Treating*, Vol. 18, 1930, pages 440-458.



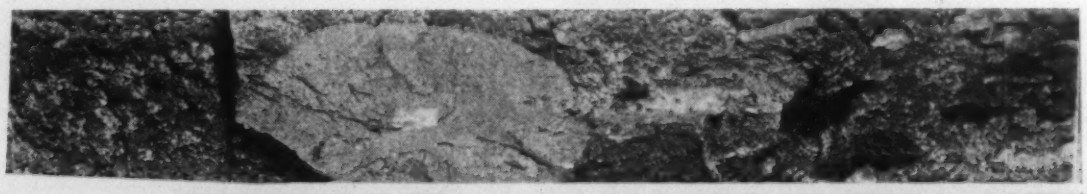


Fig. 1 (upper left). Fracture of a Crank. Composition: unknown. Macrostructure: very fine cracks visible only after etching, plate-shaped failures (first type flakes). Primary microstructure: elongated in forging direction; fracture through the matrix between the dendrites, as well as in the direction of elongation. Secondary microstructure: Martensite-like structure. Actual size.

Fig. 2 (upper right). Disk from an Ingot. Composition: C .31 %, Ni 3.35 %, Cr .81 %. Macrostructure: plate-shaped fracture near cracks (first type flakes). Primary microstructure: cracks in the matrix accompanied by long dentritic needles. Secondary microstructure: ferrite and poorly developed pearlite. Actual size.

Fig. 3 (left). Forged crank of the same heat as ingot disk shown in Fig. 2. Composition: See Fig. 2. Macrostructure: coarse grained flakes (second type flakes). Primary microstructure: as in ingot disk of Fig. 4 but not so clear as in this one due to working. Secondary microstructure: areas of ferrite and sorbite. Actual size.

Fig. 4. Disk from an ingot. Primary structure. See legend to Fig. 2. 6.5 X.

TABLE I. SURVEY OF LITERATURE *

Ref. No.	Author	Steel in which flakes occurred	Composition	Slag or oxide inclusions	Accumulation of ferrite along the cracks	Overheated structure	Higher content of alloying elements in flakes	Irregular structure due to crystal segregation	Flakes in ingot	Absence of cracks in outer parts	Coarse primary structure	Ar_3 unfavorably located	Form of flakes	Origin of flakes
1	Pacher	Ni steels	Ni : 3.20-4.10% C : .32-.35%	observed	———	———	———	———	———	———	———	———	———	accumulation of slag
2	Styri	Ni steel high C-Cr steel	———	observed	———	———	———	———	perhaps	———	———	———	———	accumulation of slag as predominent cause
3	Thum	———	———	observed	———	———	———	———	———	———	———	———	———	accumulation of slag
4	Schleicher	Ni-Cr steels	C : .25-.40% Cr : .35-.70% Ni : 3.20-3.70% Mn : .60-.75%	observed	———	———	———	———	———	———	———	———	———	accumulation of oxides due to poor deoxidation
5	Giolitti	Low alloy steels	———	observed	observed	———	———	———	———	———	———	———	Intercrystalline cracks	accumulation of ferrite. C of Fe_3C reduced by oxidizing effect of slag inclusions, ferrite thus accumulated
6	Miller	———	———	submicroscopic slag inclusions	observed	———	———	———	———	———	———	———	———	holds the same opinion as (5) but doubts that small amounts of slag inclusions have a strongly oxidizing effect.
7	Sommer & Rapatz	———	———	observed	observed	———	———	———	———	———	———	———	———	accumulation of ferrite caused by slag
8	Clayton & collaborators	gun forgings of basic electric or basic open hearth steel	C : .38% Ni : 2.9%	no statement	no statement	observed	found by micro-analysis	———	———	———	———	———	———	due to overheating (forging at too high temperatures) local melting of C and alloy rich areas
9	Rawdon Howe	Ni steel Ni-Cr steel	C : .40% Ni : 3.5%	observed	not observed	not observed	found by etching of primary structure	not observed	observed	observed	———	———	Intercrystalline fracture	Rawdon, Howe & Crouse hold the opinion that slag inclusions are not the prime cause of flakes, but cracks due to shrinkage and stresses. According to Rawdon & Howe the flakes are formed along the coarse dendrites in the ingot. According to Rawdon slag is frequently to be found in the direction of these dendrites, but these inclusions are of minor importance for the formation of flakes.
10	Crouse	Ni steel forgings	around 3% Ni	———	———	———	———	———	observed	observed	———	———	Intercrystalline fracture	
11	Hultgren	Cr steel for ball and roller bearings, acid open hearth steel	C : 1.0% Cr : 1.5% Mn : .25-.35% Si : .25-.35%	———	———	———	———	———	possible	observed	———	———	———	stresses set up during cooling subsequent to forging; inclusions may have a bearing
12	Bardenheuer	Ni-Cr gun steel, basic open hearth steel	Ni : 2.89-2.58% Cr : .57-.35% C : .31-.34% Mn : .93-.95% Si : .20-.28%	observed	not observed	not observed	found by etching of primary structure	alternatingly martensitic and pearlitic areas	not observed but author thinks it probable	observed	observed	———	No statement	cracks formed at Ar_3 due to the non-uniform shrinkage of various differently alloyed structural constituents. Slag may be a promoting cause but not a primary one.
13	Aichholzer	Cr steel	———	observed	not observed	not observed	found by etching of primary structure	eutectic carbide in hypereutectic material	observed	observed	observed	———	Coarse fracture due to hot cracks	
14	Benedicks	———	Cr : .68-1.5% Ni : 1.63-4.37% C : .13-.55% Mn : .36-.66% Si : .21-.28% one steel containing .18% Mo, another one .19% Mo, .9% W	no statement	no statement	no statement	opinion of authors (12) and (13)	explained as a typical special case	appearance not absolutely necessary	observed	known fact	found by other authors	No statement	cracks formed at Ar_3 in ingots or forgings due to non-uniform temperature distribution. Cracks in the dendrites due to poor cohesion on account of the highly alloyed outer zones of the dendrites.
15	Eilender & Kriessler	Ni-Cr (Mo, W) structural steels, 5 brands	Cr : .48-1.28% Ni : 3.05-3.63% C : .31-.42%	rare	no statement	no statement	practically always apparent	observed	no statement	no statement	no statement	location of Ar_3 and size of internal stresses in relation to deformation discussed	observed in round and square billets 100-200 mm.	crystal segregation and stresses at Ar_1 due to differences of temperatures and deformation. Secondary brittleness range.
16	Sauerwald & collaborators	Ni-Cr steels of various origin	———	possible	secondary cause	without effect	observed	possible	observed	possible	possible	possible	Flakes of first and second type	See this paper

*Opinions advanced during discussion of papers are not all included. See for instance Hultgren (11).

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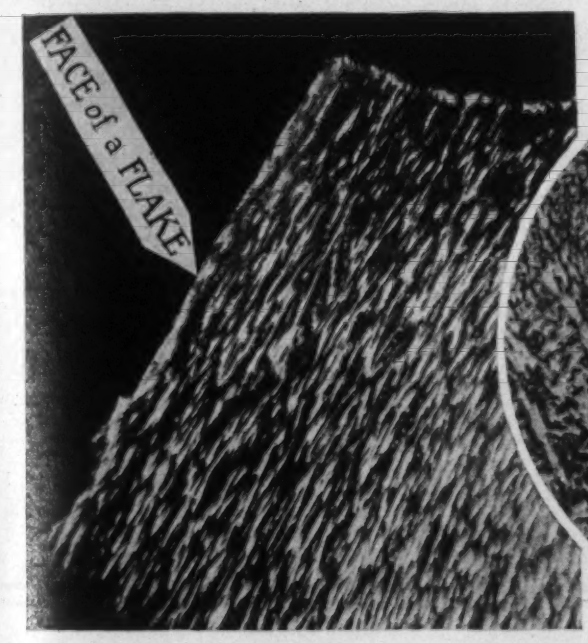

Fig. 5. Fracture of a crank. (Primary structure.) See legend to Fig. 1. 6.5 X.

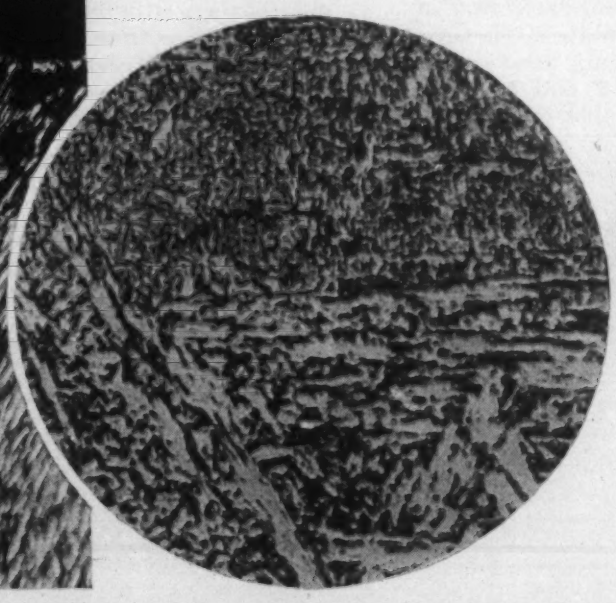

Fig. 6. Disk of a rolled round bar. (Secondary structure.) Composition: unknown. Macrostructure: very fine cracks visible only after strong etch. Plate-shaped failures. Primary microstructure: elongated in rolling direction. Fracture through matrix between the dendrites, as well as in the direction of elongation. Secondary microstructure: ferrite and poorly developed pearlite. 500 X.

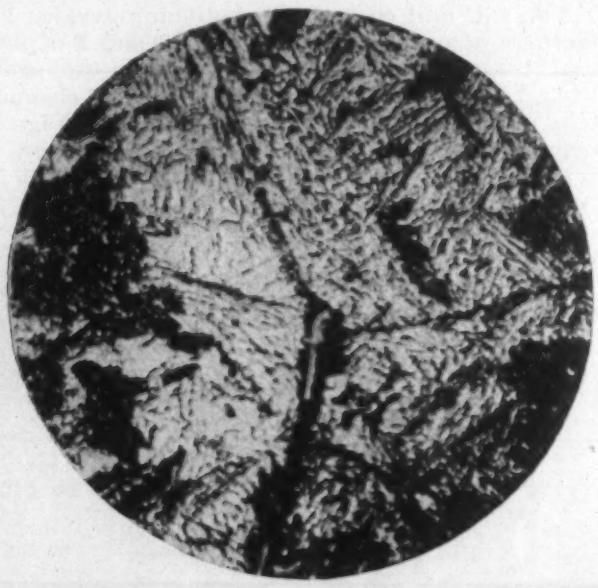

Fig. 7. Microstructure of the crank of Fig. 1 after remelting. Martensite and troostite. 800 X.

direction of the elongated dendrites (see Fig. 5). In our experiments we found this relation between segregation and flakes verified in that the secondary structure of a piece (Fig. 6) varied with the non-uniform distribution of the alloying constituents. In remelting such a piece martensite was found along the grain boundaries and troostite in the interior of the grains thus indicating crystal segregation (Fig. 7). Non-metallic inclusions may also be found in the dendrites freezing last. Such inclusions may influence the formation of flakes but they need not be a primary cause. The occurrences just mentioned apparently point to the enrichment with alloying elements of the last dendrites to freeze as the primary cause of flakes.

2. Further factors, however, are necessary to form flakes at such spots. Stresses of manifold nature, thermal stresses, rolling or forging stresses, stresses due to volume changes, for instance, hardening stresses are examples. Any of these stresses may be effective, provided they act at a low temperature at which the capacity for deformation is already low, i.e. in the range below the transformation points. There is no clear evidence yet as to the different nature of flakes of the above mentioned first and second type. It may be that the faces of a flake, apparent in the ingot, which have not been welded upon working are changed so as to show a different appearance from flakes actually formed during working or cooling.

These reasons for the appearance of flakes point the way to their avoidance. In the first place crystal segregation is to be avoided during solidification to the largest possible extent. Segregation may be equalized by annealing and forging at high temperatures. It must be borne in mind that the diffusion velocities of Cr, Ni and similar constituents of structural steels are not very high so that success depends on the circumstances.

In the second place internal stresses are to be avoided during working and cooling. Slow cooling, as some authors have shown, counteracts the formation of flakes. If, however, flakes are already present, it is possible, in principle, to weld them by hot forging. In this case, also, success depends on the individual conditions, the content of alloying elements, the possible presence of non-metallic inclusions, etc. From these facts come the contradictory statements in the literature as to whether or not flakes can be avoided altogether, or regenerated when present.

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READERS' COMMENTS

Polishing Metallographic Specimens of Cast Iron

Editor, METALS & ALLOYS:

In view of the fact that we are dealing exclusively with cast iron and particularly cast iron with a comparatively high total carbon, we have been very much interested in the various methods which have been reported lately covering the polishing of metallographic specimens. We are acquainted with the polishing procedure developed by Messrs. R. E. Schneidewind and Armando Di Giulio at the University of Michigan, having spent some time there recently under their tutelage.

The article entitled "Improved Method of Polishing Metallographic Specimens of Cast Iron," by J. R. Vilella, which appeared in the September 1932 issue of your magazine, was read with great interest, and also the comments by Mr. Samuel Epstein. It has occurred to us that perhaps it might be of interest to workers in the ordinary run of industrial laboratories to know what results we have been able to secure with cast iron specimens which were not destined to be examined at high magnifications and which had to be prepared within as short a time as possible.

You will find enclosed 4 photomicrographs. Figs. 1 and 2 are pictures of one section and Figs. 3 and 4 of another section of cast iron. All are intended to show the graphite structure. Figs. 1 and 3 show structures which we commonly used to obtain under the old method of polishing. Figs. 2 and 4 represent the type of structures which we are now able to develop. While Figs. 2 and 4 may meet with some criticism as not being entirely perfect, we feel that they are sufficiently good for our purpose and certainly are a great improvement over the structure as shown by Figs. 1 and 3.

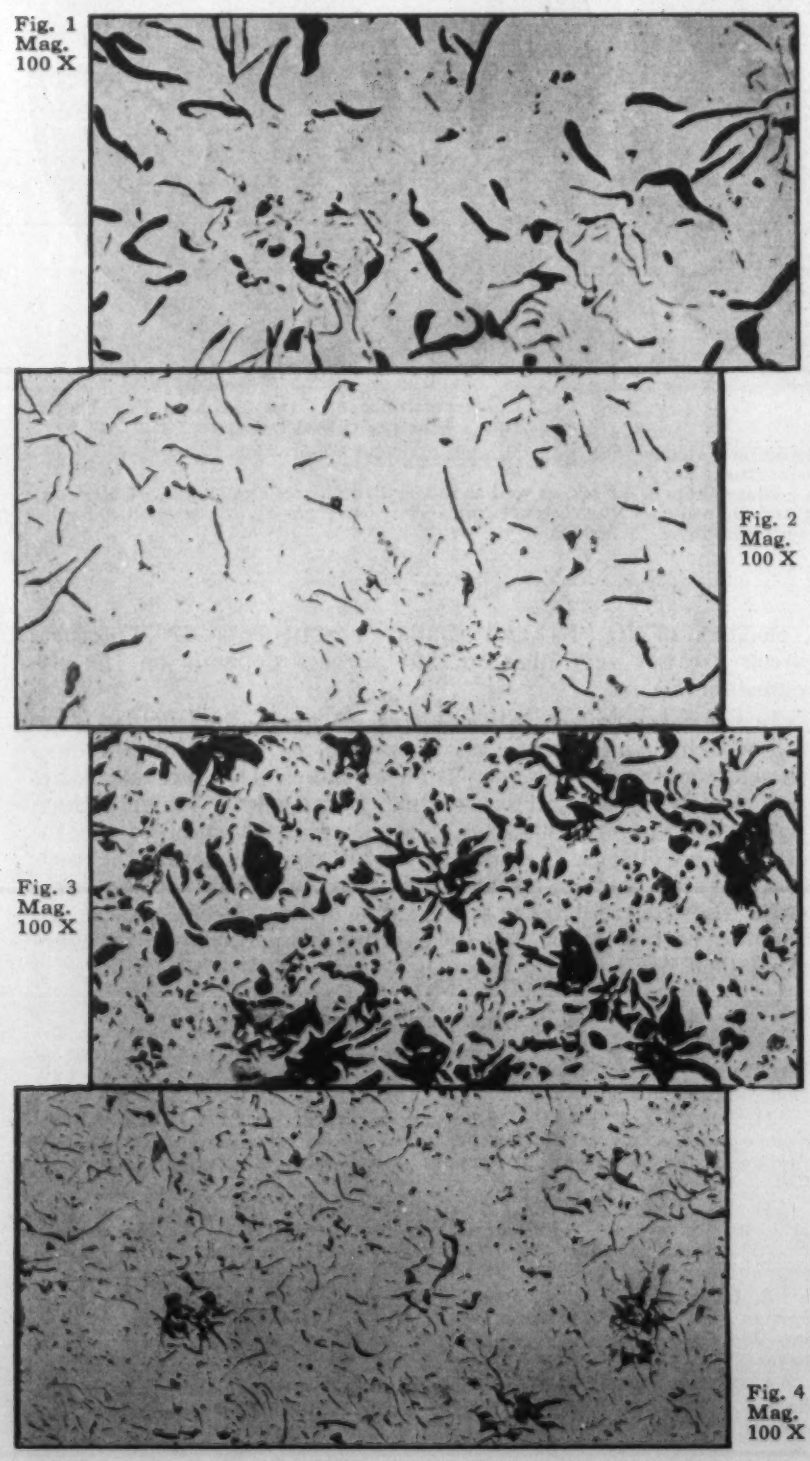

Fig. 1 Mag. 100 X

Fig. 2 Mag. 100 X

Fig. 3 Mag. 100 X

Fig. 4 Mag. 100 X

An iron such as represented by Figs. 3 and 4 is very difficult to polish without pitting. There is evidence of segregations of graphite as shown by Fig. 4 and we have found that when this occurs only extreme pains in polishing will prevent the development of a structure such as shown in Fig. 3.

The polishing procedure which we have worked out and which is satisfactory for our use is as follows:

The sample to be examined is smoothed off on a relatively fine emery wheel around 200 or 220 grit and of a soft grade. The piece is turned 90° and then the final grinding is done on a soft wheel of approximately 400 grain size.

From the grinding wheel the specimen is taken to a sheepskin buffing wheel. Chrome plating buffing compound is applied to the periphery of the sheep skin wheel and the specimen is held lightly against this part of the wheel and rotated slowly. After the grinding scratches have been removed, the specimen is taken to another sheep skin wheel which has stainless steel buffing compound on the periphery and the buffing operation is then repeated as before.

The final polishing operation is performed on a revolving horizontal wheel covered with billiard cloth. Rouge is rubbed lightly into the billiard cloth and then the loose particles brushed off. The wheel is then rotated and water is dropped on the wheel at approximately 60 drops a minute. The specimen is then either held on the wheel and rotated around in the opposite direction to the wheel travel or simply laid on the wheel, being prevented from revolving with the wheel by any suitable contrivance which will hold it in place. If the second method is used, it will be necessary to occasionally turn the piece slightly so that the familiar tails will not form.

In spite of the fact that we feel that we have a satisfactory polishing practice and one that can be completed in some 10 or 12 minutes, we are continuing to experiment and hope to be able to improve the finish on our specimens even more in the near future.

Credit for the polishing of the specimens belongs to Mr. Rudolph Flora of the company's laboratory.

Muskegon, Mich.
Feb. 22, 1933

NEIL A. MOORE
Metallurgist
Sealed Power Corporation

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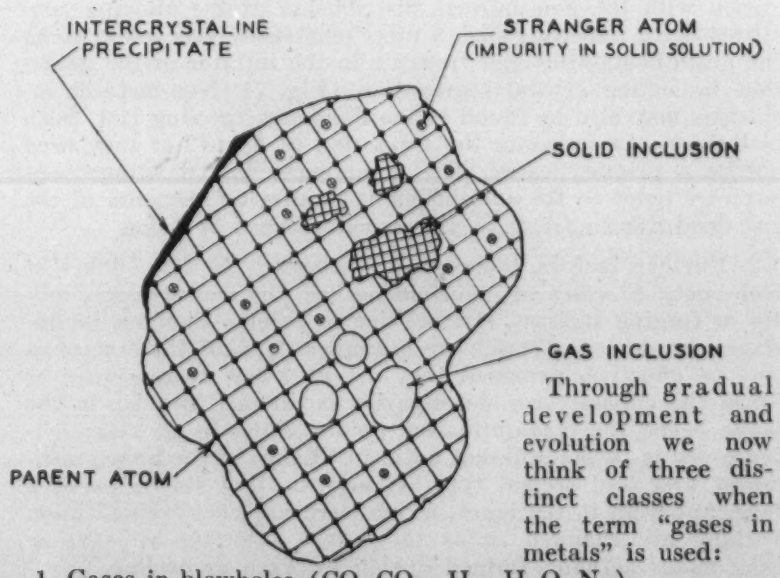
Nomenclature of Gases in Metals

Editor, METALS & ALLOYS:

The following was contributed to the discussion at the symposium on "Gases in Metals" at the A.I.M.E. meeting in February 1933, to suggest descriptive terms that are more nearly in accord with facts than the present term "gases in metals." That symposium is not to be printed, but I would like to put my convictions in regard to the matter of correct nomenclature on record.

Let us first recall the definition of a gas: a fluid having neither definite shape nor volume, and then ask what we usually have in mind when we talk about "gases in metals."



Through gradual development and evolution we now think of three distinct classes when the term "gases in metals" is used:

1. Gases in blowholes (CO, CO_2, H_2, H_2O, N_2).
2. Hydrogen, oxygen, nitrogen combined with the metals in the form of hydrides, oxides and nitrides.
3. Hydrogen, oxygen and nitrogen in solid solution in the metals.

(Continued on Page 48)

EARLY CHEMISTRY and CHEMISTS in the IRON and STEEL INDUSTRIES — No. 3

Continued from March issue

SCENES Around the Early Blast Furnaces

By L. W. SPRING and L. E. GILMORE

TO review somewhat and give a little more complete picture of the early blast furnace industry, several paragraphs from Peters' *Two Centuries of Iron Smelting in Pennsylvania* may now be quoted:

"The manufacturing of pig iron in Pennsylvania may be divided into three periods: first, the era of charcoal practice; secondly, that of development; and finally the present technical stage. The span of years from 1720 to the beginning of the use of anthracite as fuel in 1840 witnessed the period of charcoal iron making with cold blast in stone stacks with results far from being economical and operated under conditions as described in the opening pages of this monograph. The second period or era of development started with the advent of the iron stoves for heating the blast and the use of steam engines for blowing. This period saw the replacement of the stone, or brick stacks of a later date, with the iron or steel shells erected on columns, and included the beginning of the use of coke for smelting iron. The use of this fuel, coke, in a Pennsylvania blast furnace was not manifest until about 1850, although William Firmstone, whose sons for many years operated the Glendon furnaces on the Lehigh River, used coke made from Broad Top coal in the Mary Ann furnace in Huntington County in the year 1835.

"In this same period, chemists were first employed at the eastern blast furnaces. B. F. Fackenthal, Jr., then operating the Durham furnace, took a special course in chemistry at Lafayette College, while about the same time Edgar S. Cook installed a laboratory at the Warwick furnace at Pottstown.

"It was not until the dawn of the twentieth century, however, that the modern plant had its actual inception. The old-fashioned vertical hoist towers and inclined planes gave way to the skip hoist, and with this change, the single bell and hopper was replaced by the closed top with the double bell.

"Obviously it would be impossible within the narrow scope of this article to give credit to the vast majority of the blast furnace men of Pennsylvania who by their efforts have advanced the art to its present state of efficiency. As the plants grew and rebuilding became more involved, trained engineers were called into consultation to redesign and superintend the remodeling, and it seems appropriate to mention by name a few of the engineers and builders of another day, such as John Fritz, Artemus Wilhelm, William Firmstone, John Birkinbine, Benjamin Crowther, Fred Gordon, James P. Witherow, George M. McClure and Captain William R. Jones.

"With the foregoing men as the architects of the plants, their successful management was due to operators of the type of Norman Hall, Joseph Wharton, Peter Shoenberger, J. J. Spearman, George B. Wiestling, Frank Firmstone, S. Perkins, Jr., Edgar S. Cook, James Gayley, James Scott, B. F. Fackenthal, Jr., and others. Dr. Fackenthal is the only one of these men now living, and, while retired from active management of iron plants, is giving his leisure to the historic aspects of the industry. Dr. Fackenthal, who resides near the site of the old Durham Works, has made a study of old stove plates cast at the early Pennsylvania furnaces, and has been able in many cases to locate the possible place of manufacture of these castings by careful analysis of the iron from which they were made. From his knowledge of the peculiarity of the analysis of the ores which were in the neighborhood of the old furnaces, he could determine the character of iron that was produced and hence could check the probable origin of the stove plate castings.

"The blast furnace builders of today who have designed the majority of the prominent plants described in this article include David Baker, Julian Kennedy, M. A. Neeland, Frank C. Roberts, Arthur G. McKee, of Cleveland, and Freyn and Brassert, of Chicago. These plants are managed by worthy successors to the furnacemen of the old school who have supplemented their technical training with practical experience in the laboratory and at the iron notch. Leonard Peckitt, President of the Empire Steel & Iron Company, and Ambrose Diehl, of Duquesne, are mentioned to represent types of successful furnacemen of today, but a reference to the directory of the blast furnaces in the state would be necessary to give due credit to all.

Tapping a Modern Blast Furnace (Courtesy Colorado Fuel & Iron Company)

"The old iron master of Pennsylvania has gone and with him the romantic picturesque side of the pig iron industry. The furnace plants which once were the center of their individual communities have been replaced by modern types. Instead of the old hillside stack with its wooden sheds and its mule and ox teams, draying in the ore and charcoal, we have the modern furnace almost outflanked by the towering stoves and stacks, while a miniature railroad system is necessary to assemble its daily needs of raw material and to handle its product for shipment to market."

It was not until the 1880's or 1890's that many of the blast furnaces were forced to employ chemists when the change came of buying pig iron on analysis and not by fracture.

An interesting letter from Mr. G. H. Meeker, for many years Dean of the Graduate School of Medicine, University of Pennsylvania, Philadelphia, gives the early chemist's view of the blast furnace of that period.

"I was born and grew up, as it were, in the shadow of blast furnaces, in the neighborhood of Easton, Pennsylvania; in the Lehigh and Delaware Valleys; and in the neighborhood of Phillipsburg, New Jersey—in what I might call the Hibernia mines region. Later I was intimately acquainted with blast furnaces in New Jersey, in the neighborhood of New York, and in Virginia in the neighborhood of Clifton Forge. I was

actually salaried as a chemist when at Newark, New Jersey; Longdale, Virginia; and Phillipsburg, New Jersey.

"The Andover Furnace at Phillipsburg, took its name from the village of Andover, New Jersey, about fifty miles from Phillipsburg, where there occurred what was probably the most important deposit of iron ore which was ever discovered in New Jersey. It was a large and rich deposit of hematite, which was, however, simply a large pocket and became completely exhausted. The furnace plant was located at Phillipsburg mainly for traffic reasons and called the Andover Furnace; and the corporation was known as the Andover Iron Co.

"The ore was brought by the Morris Canal from Andover to Phillipsburg. The fuel was anthracite coal from the Lehigh Valley brought by the Lehigh Canal, and the flux was a dolomitic limestone which occurred at the location of the furnace. The Andover Furnace continued to operate long after the Andover deposit had been exhausted, canal transportation had ceased to be a factor and anthracite had become supplanted by coke as a furnace fuel. After the exhaustion of the Andover deposit, the rich magnetite deposits of Hibernia, New Jersey, became the source of ore; and the local dolomite continued to be used as a flux. Later some Michigan and Minnesota ores were used, and coke became the fuel.

Ruins of the Caroline Cold Blast Charcoal Furnaces erected in 1833 at Baileysburg, Pa. Capacity about four tons per day. (Courtesy Richard Peters, Jr.)

"I was well acquainted with the Andover Furnace before there was any chemical laboratory there; and, indeed, I started such work there. However, prior to the establishment of a furnace laboratory, Booth, Garrett, and Blair of Philadelphia, Pennsylvania and Porter W. Shimer of Easton, Pennsylvania, made analytical reports. I do not know when such commercial laboratory services were first used by the Andover Iron Company. However, in a discussion like this we must remember that from the earliest times the blast furnace itself really made analyses possible. Blast furnace superintendents and blast furnace 'founders' became so expert in their knowledge of their rule-of-thumb methods that in their inspection of raw materials and their daily knowledge of the behavior of the furnaces in operation, they were really doing, in a crude way, what later became the work of the chemical laboratory. In other words, they really did have their practical methods, which methods amounted to the analysis of their raw materials, of the composition of their furnace charges, and of the chemical characteristics of the pig iron which they produced.

"My first practical blast furnace chemical post was as student understudy to W. Wally Davis at Secaucus, New Jersey, in 1892. The Secaucus Furnace was located at tidewater on the Hackensack River (and on the great Hackensack Meadows) in New Jersey near New York City. It was an enterprise of the rich and philanthropic Pardee family of Hazelton, Pennsylvania. They were essentially anthracite coal operators; but through circumstances unknown to me became interested in a blast furnace at New York tidewater (Secaucus). I think the original idea was to use foreign ores landed directly at furnace; but, apparently the idea did not turn out well; and the Secaucus Furnace did not endure. Ario Pardee, the father and family head, was a great benefactor of Lafayette College, Easton, Pennsylvania, where I graduated under Edward Hart in 1893. At the time I was at Secaucus, the Superintendent was (son) Israel Pardee; and the furnace 'founder' was a very interesting, typical old-school English 'founder,' Richard (Dick) Stevens. Israel Pardee became a trustee and benefactor of Lafayette College. Stevens became the 'founder' at the Andover Furnace; and was there when I became the furnace chemist in 1895.

"It was in 1893 that I went to the New Jersey Zinc and Iron Co. in Newark, N. J. as works analyst. There were located many 'Wetherill' furnaces producing zinc oxide (mostly for pigment) directly from the oxide and silicate ores from Franklin, N. J. The Wetherill furnace residues were rich enough in iron and manganese to be melted in two small blast furnace stacks for a 'spiegeleisen' pig. When I was at Newark, my superior was George C. Stone, who had been recently promoted from analyst to metallurgical engineer.

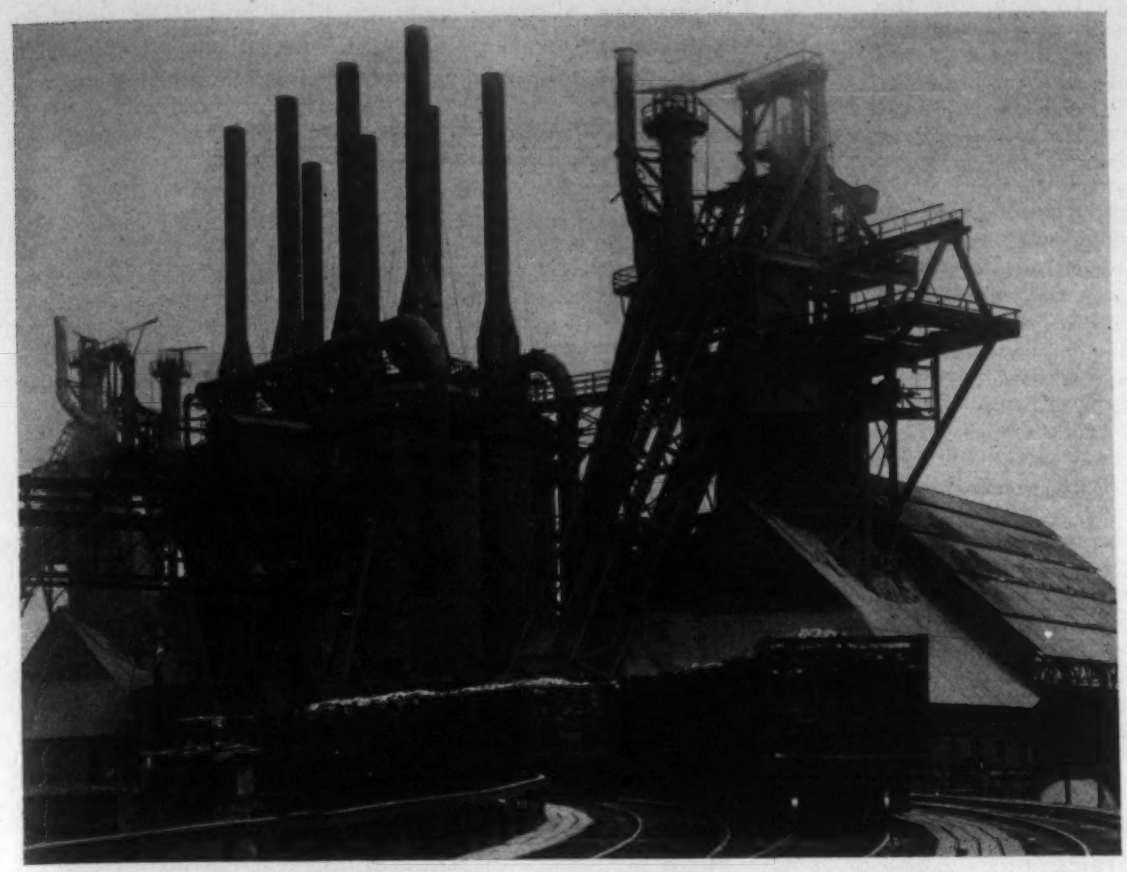
"I was chemist for the Longdale Iron Co., Longdale, Va., in 1894. I do not remember the precise dates concerned in all of the above. I finally left the iron industry when, in 1897, I took up my duties as Professor of Chemistry in the Medico-Chirurgical College of Philadelphia, and became a fixture in medical education.

"When I left Longdale, I secured as my successor there, Walter A. Harper, Lafayette, '94, who is now a public accountant in Newark, N. J. He did not remain long; and was succeeded by James C. Attix, Lafayette, '95. Attix later succeeded me at the Andover Furnace. Later he joined me as a student assistant in the Medico-Chirurgical College; and successively obtained there his D.D.S. and M.D.; and later became Professor of Chemistry in the professional departments of Temple University in Philadelphia, where he is still connected as one of their older professors.

"You will see from the dates that the changes of post were rapid in the cases stated. The post of chemist at small blast furnace plants in the late eighties and early nineties was neither important nor lucrative. Seventy-five dollars per month was common pay; and twice that was large pay. So far as my observation went, the chemists in small blast furnace plants were mostly fresh college graduates with but little experience and with duties as routine analysts. Their hope was to grow to be superintendents of small furnaces; or perhaps become connected with large steel works and grow therein. The tendency was to move about rapidly as new posts offering a little better pay, or working conditions, or prospects appeared. The furnaces were on twelve-hour shifts; and often the chemists were expected to work about twelve hours daily and some Sunday hours.

"This was done so that the superintendent could know the chemical grade of the pig and slag promptly, and act accordingly; and so that the pigs as they went from the casting beds or 'chills' could go directly on cars to fill orders or directly to chemically graded storage. Since there was usually only one chemist, if he did not work long hours there was more or less the 'Devil to pay' with the furnace operation or shipping and storage routine.

"In small furnace plants specializing in foundry pig, largely graded by 'fracture,' it was different. But this was a time when the large steel plants were buying much pig from the small furnaces which could meet the chemical specifications, especially as to sulphur. The phosphorus largely took care of itself and likewise the silicon, especially if one could also see the fracture. However, the sulphur was always a question mark from cast to cast; and this held the young chemist to his long hours. This was also about the time when the steel works began to pay a premium for 'chill' iron—pigs cast in big cast-iron pig moulds or 'chills.' The freedom from adhering sand was



Blast Furnaces in the Pittsburgh District. These furnaces are ready to produce almost a thousand tons per day. (Courtesy Carnegie Steel Company)

fine for the steel maker; but it destroyed the pig 'fracture' and helped to keep the chemist long hours in his laboratory. I don't know the present conditions; but I surmise that the present-day chemist in small blast furnace plants are better paid, more appreciated; less worked as to hours and activities; and provided with suitable assistants and better facilities.

"I can say that in those days, the laboratory was usually any old room or shack which happened to be available, because not much wanted for anything else anyway. The simplest unpainted carpentry was good enough for the chemist; and his furniture was any old surplus stuff that happened around or that the carpenter knocked together for him. Other surroundings were in keeping. No doubt the commercial laboratories would in many cases have continued much longer to get routine samples had it not been for the imperative need which had arisen for chemical grading of the pig-iron with utmost promptness. This made the chemist essential to small furnaces; and the commercial laboratories (of which there were few) were used as checks and as referees in vendor-vendee disputes as to quality and price.

"The lot of the small blast furnace chemist was an unenviable one; but the times were different and young men were then uncoddled products, who not only got little, but also expected little. The lot of the furnace laborer was on a parity. Wages of common labor at $.75 per day were well known to me; and $1.50 per day was a 'good' wage in plants I knew with twelve-hour shifts. When in these days I see young college graduates getting minor teaching posts at salaries that are not too large; but which would have been dream-like to the men of whom I have just written; and when I think of their few hours of work and months of vacations—I smile reflectively as memory of my early humble labors comes to me. Which is the better? Frankly, I am not sure of the answer; but I do feel sure that hardship is a better schoolmaster than comfort.

"I do not join in the rather general disdain with which this rule-of-thumb method has generally been regarded. On the contrary, I admired the practical skill and knowledge which the blast furnace overseers exhibited in those old days. I will give you an anecdote of the Andover Furnace group, which will illustrate, in a crude way, the fact that even the wives of the furnace men had some conception of the value of pig iron quality.

"At one time, certain of the furnace employees were paid on a sliding scale depending on the grade of pig iron produced. This was done in order to spur the employees to use their utmost skill in producing the highest grade of pig iron possible, and thus bring extra profits to themselves and to their employers alike. In those days, practically all of the heavy labor was done by Irish workers. One of these was a 'keeper' (I am assuming that my expressions, 'founder,' 'keeper,' etc. will be well understood), who followed what was the common custom of many of the Irish workers, of carrying his pay envelope home and giving it into the custody of his wife for prudent use. On this particular pay day, on his way home, the 'keeper' spent some of his wages at a local tavern, resealed his envelope and handed it to his wife. When she opened it she complained of the fact that he must have removed some of the money, which removal he denied and, in explanation, stated that the furnace had been making only grey-forge iron. Her reply was, 'Yez lie, I saw the kish on yer shir-r-t.' In other words, she knew the furnace was making high-grade iron, which meant higher wages.

"I could tell much about the hours of work and the character of the labor, etc., but it would be too time consuming. No one who has not lived around a blast furnace in the old days, and who knows only the smooth operation of the present days, can realize the 'living thing' which the furnace stack always seemed to all of the personnel who cared for it. The stack was always referred to as 'she' and 'her'; and the workmen, especially the 'keepers,' prided themselves on their skill and became locally honored for the same. Furnace chillings, scaffoldings, blow-ins, blow-outs, casts, cinder flushes, and gassings, and how 'she' was working were daily, almost hourly, matters of interest. When she chilled, blew-in, blew-out, or someone was gassed, 'she' was surrounded not only by all of those on duty who could be there, but also by many of those who were off duty, and those in the neighborhood who felt that they also knew 'her'. It was a common cry, for instance, if it was difficult to open the iron notch (long before the 'gun' was invented) for a 'council of war' to get together and decide how to get the iron notch open, and how to keep it open after the tap was started. Only certain ones were considered expert enough to (as the Irish put it) 'put-t-t the precker to her'. Many a time I have seen a half-dozen men ramming with the same 'precker' over the intense heat of the 'trough'. I will tell you of one of the curious difficulties which happened at the Andover Furnace while I was chemist there.

"The blast was off for a cast when suddenly a great flow of a very liquid material rushed in through the tuyeres, and into the belly pipes and solidified there, to such an extent as to necessitate taking down the belly pipes and cleaning them and the tuyeres before operation could be resumed. The material in question proved to be a fairly pure potassium cyanide.

"I have had many experiences with furnace cadmia which were very rich in zinc oxide; also with scaffolds, due to the titaniferous character of the ores; and spinel formations due to the dolomitic limestones which very much complicated the slag analyses.

"It was in Virginia that I saw more 'gassings' than in any other section. This was largely because of the full brick construction of the furnace stacks and the unavoidable furnace gas leakages. I remember one instance at Longdale when a workman became unconscious from furnace gas, and dropped from near the top of the furnace stack where he was working, and was saved from injury only because the seat of his trousers caught on a hook supporting the furnace hoop. There he hung, suspended between the furnace top and the ground, until a ladder crew released him and revived him.

"I remember a very amusing fake 'gassing' at the Lowmoor Furnace in Virginia. These stacks were located right beside the Chesapeake and Ohio Railway tracks. The 'founder,' at that time, was an Englishman and was quite a local celebrity; and, like many of the old 'founders,' was occasionally on duty somewhat inebriated. On this occasion he was inebriated at a time when some members of the Board of Directors were on a train which they were influential enough to have stopped right beside the furnace, in order to dismount. The 'founder's' loyal employees, knowing his condition and seeing the Directors appear, ran to him, took him on the hoist to the furnace top; and arranged it so that just as the Directors appeared on the scene, the 'founder' was brought down on the hoist and taken

away, ostensibly unconscious from 'gassing.' In this way the 'founder' was not only saved from censure or worse, but he was the object of sympathy on the part of the Directors, who, in a meeting which they held within the hour, took due note of the danger pertaining to the blast furnace work, expressed their sympathy for the 'founder,' and voted him an increase in salary. So far as I know, the Directors were never advised of the real facts concerning this particular 'gassing.'

"In Virginia the workmen were not Irish. They were Virginian mountaineers. At Longdale the plant had been in operation since very early days. Originally the furnace had used charcoal as the fuel, with local limonite and limestone. Their speech was, of course, the characteristic southern mountaineer's speech. They had no life, except the life around the furnace, in their cabins in the mountain 'hollers,' in the mines, and in the 'qurries' (Quarries), and the company store. Common labor was valued at only 75 cents a day at Longdale when I was there in 1893. Practically all food stuffs were purchased from the company store and there was little used except 'sow-belly,' corn-meal, and beans. The 'sow-belly' was a sight. It came in box car loads and was piled in the store, like brick in a well, amid the dirt, the insects and vermin, and even kerosene oil.

"In the southern speech those in authority were usually known by their first names, especially since there were comparatively few families; and the use of the first name disclosed the person intended. For example, Mr. Harry, Mr. Guy, Mr. Arch, and Mr. Jim, etc. A type of colloquy would be something like this:

"Manager: 'Good-morning, Arch.'

"Answer: 'Good-morning, Mr. Harry.'

"Manager: 'Where is Jack this morning'?

"Answer: 'Why I haint saw Jack this morning and I haint saw nobody what have saw him this morning.'

"Many blast furnaces which were well known to me are now only memories, but to many of the folks who knew them they remain as fond memories, and memories surrounded by many things which would seem quaint in these days. I think of the Glendon Lucy and Durhamneer Easton, and the Stanhope Furnace in New Jersey. Even today the folks living along the Delaware river refer to certain flat bottom boats as Durham boats. The reason is that in those earlier days at the furnace these flat bottom boats were built during the low water times on the Delaware river, where the pig iron was simply stored. Later on, when the rush of the high water occurred, these Durham boats were loaded with pig iron, and floated to tidewater; the boats were then sold for what they would bring. New boats were then built for the next high water.

"If you wish to know of blast furnace men, who in those earlier days were celebrities in the Lehigh Valley and New Jersey districts, you should ask concerning the Fackenthals and the Firmstones and their associates. Personally, I have been out of association with these things for so long that I cannot be of much help other than the information which I have given above. I have had no association with the iron industry since 1897, more than 34 years ago, and the men and technology have largely gone on and left me behind. The cause for this is that the time came in medical education when the medical schools desired to have professional chemists to teach chemistry to students, instead of physicians who only had chemistry as a side issue. From the time I began as a Professor of Chemistry in the Medico-Chirurgical College in Philadelphia in 1897, I have remained intimately connected with medical education in America; in which I have been playing a rather conspicious part, especially since I have been a pioneer in the development of graduate medical education in America upon a definite basis."

(Continued in April Issue)

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(Editorial—Continued from page A9)

detected. Similar cases have been noted by Evans and others where galvanized iron has acquired a highly protective skin under corrosion. These, however, are unusual cases. Evidently much depends on the condition of initial exposure as illustrated by the work of Vernon, Bengough, Evans and others on protective films. If the environment is favorable to the formation of a dense adherent protective layer, striking results may be obtained even with ordinary wrought iron or steel which are admittedly vulnerable under some conditions. Therefore, conclusions should not be drawn from isolated cases of unusual life of any of these metals.

The nature of metal surface films and their bond to the metal might well be a profitable subject of scientific research on a coöperative basis.—F. N. SPELLER

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(Readers' Comments—Continued from page 44)

Among these three classes the substances in the first and the third meet the definition of gases, while those in the second certainly do not. They are solids with definite shape and volume in the temperature range with which we are concerned.

Furthermore, in order to arrive at a logical classification we should include in Class 3 the other interstitial impurities in true solid solution such as carbon, sulphur, and phosphorus, as in this form they are gases just as much as uncombined hydrogen, oxygen and nitrogen.

The term "gases in metals" was originally used only in connection with gases in blowholes, and I think it would be well if we should limit it to this class. It was extended to cover the other two classes chiefly because of the methods developed for determining the impurities in metals, by which the impurities were obtained in gaseous form (oxygen as CO, CO_2 or H_2O, hydrogen as H_2 or H_2O, nitrogen as N_2). The only reason carbon has not been included in the term "gases in metals" is that we usually do not think of carbon as a gas. But in the metals oxygen occurs as atomic oxygen and as oxides, hydrogen as atomic hydrogen and hydrides, nitrogen as atomic nitrogen and as nitrides, carbon as atomic carbon and as carbides, and why not add sulphur as atomic sulphur and as sulphides, phosphorus as atomic phosphorus and phosphides. They are all in the same class structurally while in the metals. Why should we separate them into different classes because of their characteristics outside the metals with which their effect on the metals has nothing to do?

Oxygen is essential to the formation of blowholes, but oxygen by itself will not form a blowhole, at least not in base metals. Carbon is just as essential.

Oxygen, nitrogen and hydrogen have determining effects on the physical properties of metals, but so do carbon, sulphur and phosphorus.

We thus see that there is no logical reason for the term "gases in metals" as we use it today and the sooner we decide to limit it to its proper sphere the more confusion we shall save ourselves in the future.

What term then should we use? I can think of no one term that can logically be applied to the three classes listed at the outset, but I see no reason why we should need one term to cover them all. In my opinion we should carefully keep them separate and distinct because of the effect they have on properties of the metals. For this reason I would suggest the following classification of impurities in metals for consideration:

1. *Inclusions.*
   (a) Gaseous inclusions (gases in blowholes).
   (b) Solid inclusions (carbides, nitrides, oxides, sulphides, phosphides, hydrides and graphite).
2. Impurities in Solid Solution (C, O, H, N, S, P in atomic dispersion).

In conclusion may I emphasize the fact that blowholes are inclusions or precipitates just as much, and in the same sense, as carbides, nitrides, etc.; also that a compound is a separate phase and consequently should be regarded as a precipitate or inclusion, no matter how small the particle; and finally, that impurities in solid solution, while present in very small amounts, should be considered separately on account of their great effect on the properties of metals.

East Pittsburgh, Pa.
March 1, 1933

T. D. YENSEN,
Research Laboratories,
Westinghouse Elec. & Mfg. Co.

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Summer Course in Practical Spectroscopy

The Massachusetts Institute of Technology is offering a course in practical spectroscopy during the six weeks from June 13 to July 25, which will deal largely with applications of spectroscopy to biology, chemistry, geology, metallurgy and other branches of applied science. A research conference on problems in spectroscopic analysis is being planned for the last two weeks of the course, when manufacturers of spectroscopic equipment designed for analytical work will have apparatus on display and in operation.

At the same time a laboratory course in applied spectroscopy and one in structures of atoms and molecules based on spectroscopic evidence will be offered.

The application of thermodynamic reasoning to the study of constitutional diagrams is not new, for it had been attempted in the early days of metallography, particularly by Ruer, van Rijn, Roozeboom, and Tammann. But until recently the treatment had been purely formal, chiefly because of lack of data on specific heats, heats of transformation, heats of fusion, etc. Now, however, it seems possible to treat metal systems quantitatively, and this has been done in the present article for the important Fe-C system. It may confidently be predicted that this sort of study will shortly become an integral part of metallurgical research.

The importance of thermodynamic treatment of constitutional diagrams—aside from the familiar use of the phase rule—lies in the information which it furnishes concerning the molecular and atomic constitution of the phases concerned. Such data, for this and other systems, are of the greatest importance, first with respect to the understanding of constitutional diagrams and of the behavior of the phases in solid-solid transitions, and second with respect to the reactions in which these phases participate, in both solid and liquid state, with gases, slags, etc. Thus Körber and Oelsen show that C dissolved in liquid Fe is dispersed in the form of molecules of Fe_3C, with only a slight dissociation, though because of insufficiently accurate data they were unable to decide whether, from the thermodynamic point of view, C in austenite is dispersed as atoms of C or molecules of Fe_3C. Besides reviewing this work critically Dr. Austin has performed a service in rendering the formulas into the symbols used in this country. The very complete data on the constitution of the Fe-C system which Körber and Oelsen have assembled show many points of discrepancy, particularly with respect to the position of the A_{cm} line. Even upon such an old problem as that of the constitution of Fe-C alloys, new and precise data would be very welcome.

A Thermodynamic Study of the IRON-CARBON Diagram

An Extended Abstract

By J. B. AUSTIN*

based on a paper† by Friedrich Körber and Willy Oelsen****

THE PRIMARY function of the so-called equilibrium diagram is to represent as a function of temperature the equilibrium relations between the various phases which can exist in a given system. The common and most satisfactory way of constructing these diagrams is to make direct observations by means of any one of a number of well-established methods such as the microscopic examination of quenched specimens, or thermal analysis. It is also possible to fix the positions of some of the lines in the diagram, at least semi-quantitatively, by calculations based on thermochemical data, if such are available. This method, like most indirect methods, is not very satisfactory when used by itself since the results are rather sensitive to errors in the thermal data. It is, however, of great value in a critical examination of the existing data. It frequently enables the selection of the best data from among those reported in the literature or the detection of inconsistencies in the results of a single investigator and it occasionally permits the drawing of conclusions regarding the nature of the solid or liquid solutions observed experimentally.

The successful application of the thermodynamic method depends on 2 factors: first, the existence of sufficient data of requisite accuracy obtained under conditions of true equilibrium, and second, a discriminating use of the equations themselves. It cannot be too strongly emphasized that one can never draw a complete equilibrium diagram from thermodynamic considerations alone. The equations employed are essentially a set of machines into which certain experimental results are fed and from which, after turning the crank properly, information regarding the equilibrium diagram can be obtained. Not only is it necessary to have the numerical data to feed into the machine but in some cases the general outline of the diagram must also be known since it determines which of the machines or equations must be used. It is obvious that the accuracy of the computed results can never be better than that of the original data, and it is not infrequently poorer.

Because of the great practical importance of the iron-carbon alloys a number of attempts have been made to treat them thermodynamically. These studies have in general been rather unsuccessful because the thermal data for this system are far from satisfactory and the thermodynamic equations have not always been applied with due regard to their limitations. However, there are a few discussions which have contributed something of value. Among these, the paper by Körber and Oelsen, which is the subject of this abstract, is one of the best because it is marked by an unusually complete application of thermodynamic principles. For this reason and also because their discussion leads to some interesting conclusions it has been considered worth while to present the more important features of their work in English.

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†Thermodynamische Betrachtungen zu einigen Gleichgewichtskurven des Zustandsschaubildes Eisen-Kohlenstoff. *Archiv für das Eisenhüttenwesen*, Vol. 5, 1931, pages 569-578.

**Kaiser-Wilhelm-Institut für Eisenforschung, Düsseldorf.

I. Thermodynamic equations for the equilibrium lines in a binary diagram.***

We have to consider the general case of a substance A, which undergoes a change of phase at temperature T_t passing from phase a to phase a'. This phase change can be a boiling point, a melting point or a crystallographic inversion point since thermodynamics does not distinguish between these three; it cannot be a Curie point or magnetic inversion. The equations describing the way in which the transformation temperature is affected by the addition of another substance B are well known and can be found in any standard text. In order to avoid confusion, however, Körber and Oelsen discuss in some detail the equations which they use and the assumptions upon which these equations rest. The discussion is divided into several sections each based on a different set of assumptions regarding the nature of the solubility relations between the 2 components.

Case 1. The second substance B dissolves in the high-temperature phase (a') but is insoluble in the low temperature form (a). The change in T_t for this case can be found from the equation

$$\frac{d \ln N_2}{d T} = \frac{\Delta H}{R T_t^2} \qquad (1)$$

where N_2 is the mol fraction‡ of A in the solid solution a'_B, R is the gas constant (1.989 cal. per deg.), T_t is the transformation temperature on the absolute scale and ΔH is the differential heat of solution, that is, the heat of solution of one gram-atomic weight of A in the form of phase a in a very large amount of saturated solution of B in a' (which we shall denote by a'_B). For dilute solutions ΔH is approximately equal to the latent heat of the phase change in pure A, that is, to the heat of melting, heat of transformation, etc.

***The notation and conventions employed in the original paper have been changed to conform with those of Lewis and Randall (Thermodynamics, McGraw-Hill, New York, 1923) which are the ones in common use in this country.

‡The mol fraction is a means of expressing concentration. It is the ratio of the number of gram-atoms (or gram-molecules for a compound) of the substance A to the total number of gram-atoms (or gram-molecules) of all the components in a system. For example, in this case it is equal to the number of gram-atom weights of iron divided by the sum of the number of gram-atom weights of iron and the number of gram-atom weights of carbon. If one considers the solute to be Fe_3C rather than C the number of gram-formula weights of Fe_3C is used instead of the gram-atoms of C.

It is frequently more convenient to use this equation in an integrated form and the integration is easily performed if it is assumed that A and B do not form a compound, and that ΔH is independent of temperature and of the concentration of the saturated solution. With these simplifications we get

$$\ln N_2 = -\int_T^{T_t} \frac{\Delta H}{RT^2}\,dT + \ln N_t = -\frac{\Delta H}{R}\left(\frac{1}{T} - \frac{1}{T_t}\right) \quad (2)$$

where N_t is the mol fraction of A when the phase change occurs at T_t. Obviously N_t is 1 and $\ln N_t = 0$. The heat effect, ΔH, is now the sum of the molar heat of transformation of pure A and the heat of solution of 1 mol of A in the form of phase a' in a large amount of saturated solution a' . The latter effect is usually small compared to the former and it is commonly assumed that ΔH equals the latent heat of the phase change, $a \rightleftarrows a'$.

If the phase change is a melting, the assumptions mentioned are substantially fulfilled; if it is a crystallographic inversion, some error is usually introduced. For the latter ΔH should be expressed as a function of T, thus,

$$\frac{d \ln N_2}{dT} = \frac{f(T)}{RT^2} \quad (3)$$

or

$$\ln N_2 = -\frac{1}{R}\int_T^{T_t} \frac{f(T)}{T^2}\,dT \quad (4)$$

Case 2. The substance B is appreciably soluble in both phases of A. In this case we have

$$\frac{d \ln \frac{N_2'}{N_2}}{dT} = \frac{\Delta H}{R\,T_t^2} \quad (5)$$

where N_2' is the mol fraction of A in the saturated solution of B in a' (a'_B), N_2 is the mol fraction of A in the saturated solution of B in A (a_B), and ΔH is the heat absorbed when one mol of A is transferred from a large amount of a'_B to a large amount of a_B. This heat effect is composed of 2 parts, the heat of transformation of pure A at the temperature T_t and the algebraic difference in the heat of solution of A in a'_B and a_B. If this difference is small ΔH is equal to the heat of transformation ($a \rightleftarrows a'$) and this assumption is commonly made in applying the equation.

The integrated form of equation (5) is

$$\ln \frac{N_2'}{N_2} = -\frac{\Delta H}{R}\left(\frac{1}{T} - \frac{1}{T_t}\right) \quad (6)$$

which rests on the same assumptions as are involved in equation (2).

Case 3. In addition to the lines representing a change of phase the diagram also contains lines which are saturation curves, for example, the lines QP and SE on Fig. 4. If the solute separating from the solid solution is practically a pure component, equation (2) can be applied by converting

$$\frac{\Delta H}{RT_t}$$

into an empirical constant. We then have

$$\ln N_2 = -\frac{\Delta H}{RT} + \text{const.} \quad (7)$$

In this case ΔH is a real heat of solution which cannot be predicted or even approximated with any certainty from any property of the pure components.

Equations 1 to 7 are sufficient to cover all the important lines in the iron-carbon diagram. It should be emphasized once more that the results secured by the application of these equations are valid only to the extent that the assumptions involved in the derivations, including the integration of the differential equations, are fulfilled; if approximations are introduced in performing any mathematical operation or in estimating the numerical value of any datum required, the results may be appreciably influenced.

II. Calculation of the lower boundary of the austenite field, the so-called A_3 line (line GOS in Fig. 4).

The solubility of carbon in ferrite being relatively very small equation (4) is applicable if ΔH is expressed as a function of T. The introduction of $f(T)$ in place of a constant ΔH constitutes an advance since, so far as the abstractor is aware, this is the first time that the variation of ΔH with temperature has been taken into account in considering the Fe-C system.

The function of T is evidently not a simple one, a circumstance which makes analytic integration rather difficult, so that the integration is more easily performed graphically. For this purpose it is convenient to change the form of the equation slightly by substituting mol percent carbon (C') for the mol fraction (N_2) and by using common logarithms in place of the natural ones. Making these changes we get

$$\log C' = -\frac{1}{2.303R}\int_T^{T_t} \frac{f(T)}{T^2}\,dT + 2 = +\frac{1}{2.303R}\int_T^{T_t} (H\gamma - H\alpha)_T\, d\left(\frac{1}{T}\right) + 2 \quad (8)$$

The term $(H\gamma - H\alpha)_T$ represents the difference in heat content of α iron and of β iron at a temperature T. This quantity can be obtained directly from the experimental data on total heat by plotting the heat contents of the 2 forms, extrapolating the curve for γ iron below the A_3 point and then reading off the difference at various temperatures. If this difference is plotted against $1/T$, the area under the curve between $\frac{1}{T_t}$ and $1/T$, which can be measured with a planimeter, corresponds to the integral $\int_T^{T_t} \frac{f(T)}{T^2}\,dT$. To obtain the first term on the right hand side of equation (8) it is only necessary to multiply this integral by $1/2.303\,R$ ($= 0.218$).

The numerical results for the integration using the total heat data of Oberhoffer and Grosse[1] are shown in Table I. Similar calculations were also made using the data of Wüst, Meuthen and Durrer[2], of Umino,[3] and of Klinkhardt[4] and the results are shown in Fig. 1. The solid line represents the mean for

[1] Oberhoffer and Grosse. *Stahl und Eisen*, Vol. 47, 1927, page 576.
[2] Wüst, Meuthen and Durrer. *Forschungsarbeiten auf dem Gebiete des Ingenieurwesens*, 1918, page 204.

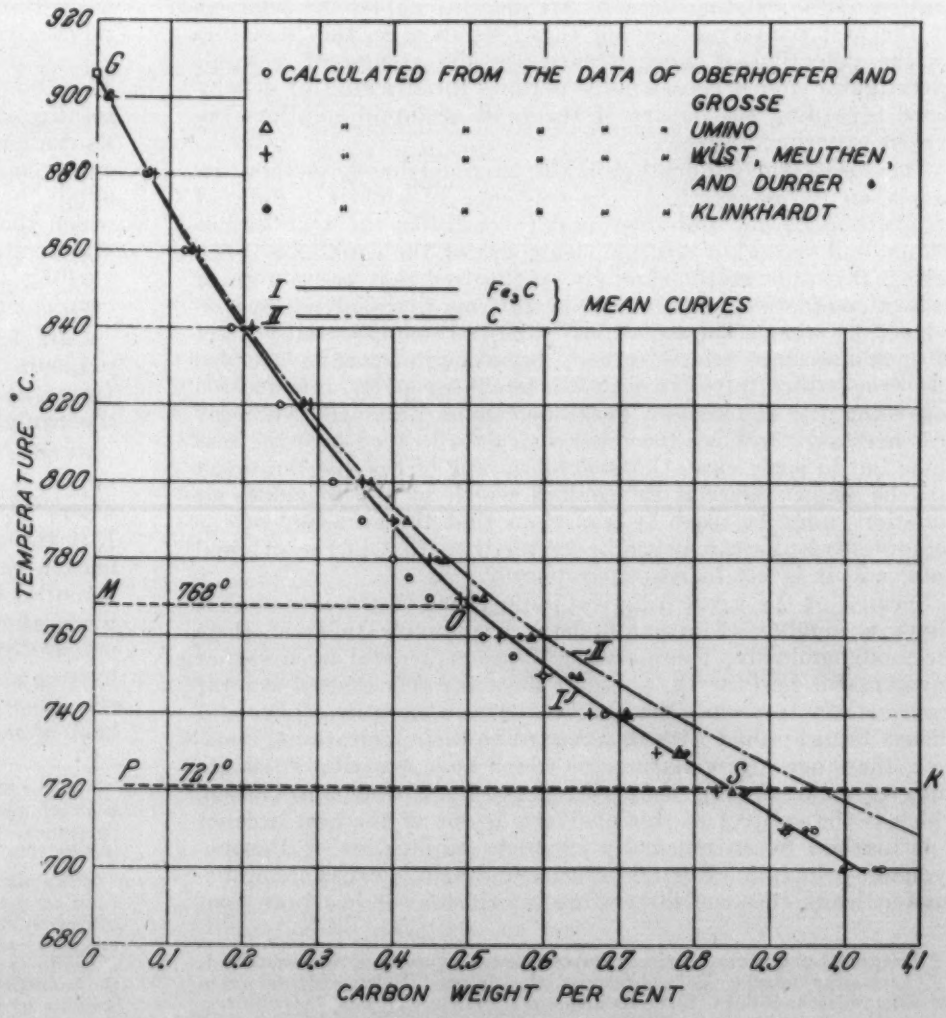

Fig. 1. A_3 line calculated from the heat content data.

the several calculations based on the assumption that the carbon is present as Fe_3C; the assumption that the carbon is in solution as free carbon leads to the dot and dash line (II).

The authors call attention to the fact that in spite of the rather divergent results obtained by the various investigators for the total heat the lines calculated from the several sets of data are in remarkably good agreement. The abstractor would like to point out that this agreement is not all that it appears to be at first glance since the use of the total heat data above and below T_t introduces but a small correction to the heat of transformation which is the important factor, and the authors appear to have used practically the same latent heat in all their calculations; for example, it is specifically stated that in the case of Klinkhardt's work, they did not use his value of ΔH (216 cal./g.-atom) but merely employed his specific heat data in conjunction with Oberhoffer and Grosse's value of 325 cal./g.-atom. It is to be expected then that the calculations should lead to lines which are in agreement. If one uses Klinkhardt's value for ΔH as well as his specific heat data one obtains points which lie considerably below any shown in Fig. 1; for instance, at 840°C. the equilibrium percentage of C is approximately 0.11 while at 800°C. it is 0.25. The agreement obtained by the authors is, therefore, directly the result of their selection of the data.

Any changes of this sort, while influencing the actual position of the line, do not alter certain general conclusions which can be drawn. The line GOS is certainly convex to the concentration axis and the curvature appears to be greatest at approximately 768°C. Moreover, the line is continuous and shows no break at this temperature such as would be expected if β iron were a true phase. It is concluded, therefore, that non-magnetic α iron is not a separate phase, a conclusion which is in agreement with evidence from other sources.

In Fig. 2 the mean curves from Fig. 1 are compared with the experimental determinations of the line GOS. The agreement is on the whole good and curve I cuts the 721° coordinate at 0.86% C. which is the composition generally accepted for the eutectoid point. It should be pointed out, however, that if one uses the heat of transformation reported by Klinkhardt the calculated curve (not shown) lies somewhat lower than I falling almost on the experimental points of Sato. The agreement between curve II, which represents the mean line computed on the assumption that the carbon in austenite is in the form of free carbon, and the observations are not so good as for curve I. This indicates that the carbon is largely in solution as carbide, but the uncertainty of the heat content data and of the experimental observations is so great that no definite conclusion can be drawn. Line III is a curve drawn through the points of Maurer who cooled his specimens quite rapidly; it has no bearing on the thermodynamic argument.

In the authors' opinion the fact that most of the experimental points lie on or under curve I confirms the observation that C has a very small solubility in ferrite since an appreciable solubility would have resulted in the calculated line falling above the observed points. In view of the way in which curve I was computed, however, it seems to the abstractor that this argument is not of great significance.

[3] Umino. *Science Reports, Tôhoku Imperial University,* Vol. 15, 1926, page 597; Vol. 18, 1929, page 91.
[4] Klinkhardt. *Annalen der Physik,* Vol. 84, 1927, page 1672.

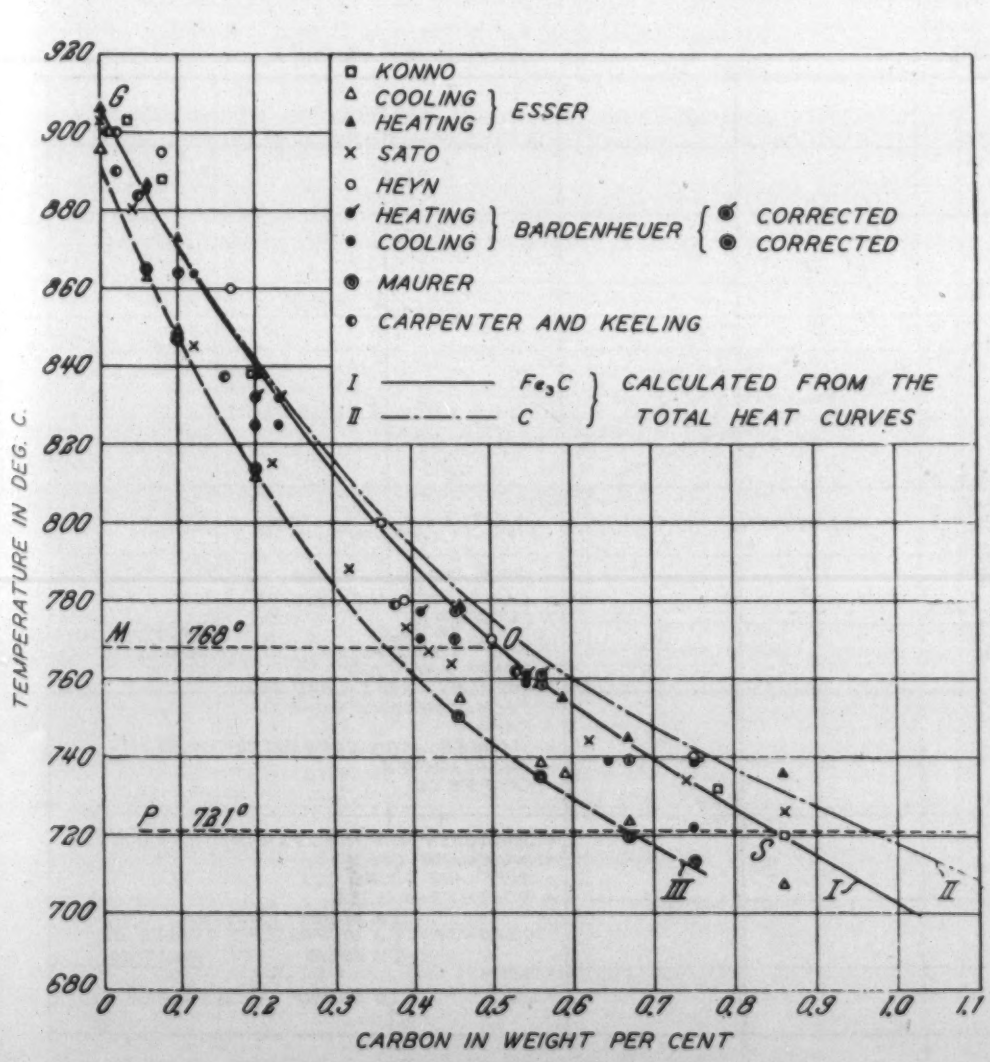

Fig. 2. Comparison of the thermodynamically calculated A_3 line with the experimental points.

III. The form of the saturation curve for Fe_3C in austenite, the so-called A_{cm} line (line ES in Fig. 4)

Since the solubility of γ-iron in Fe_3C is very small and since the equilibrium carbon content of cementite crystals is practically constant with temperature, equation (7) is applicable. While the experimental observations on the course of the line are not at all concordant there is substantial agreement on the position of the terminal points. The lower end cannot be far from 0.86% C at 721°C. while the upper end must be close to 1.76% C at 1145°C. Using these data the constants in equation (7) can be evaluated and we obtain

$$\log C\gamma = \frac{-1188}{T} + 1.839 \qquad (9)$$

where $C\gamma$ is the total carbon content of the austenite saturated with C expressed as mol percent Fe_3C.

When calculated to weight percent of carbon the curve proves to be practically a straight line which is almost certainly as accurate as most of the experimental data (Fig. 4). The slope of this line represents the heat of solution of 1 mol of Fe_3C in a large quantity of austenite. Thus $\Delta H = +1188 \times 2.303R = +5{,}440$ cal per mol or + 30.2 cal per g. of Fe_3C. Since the assumption that the carbide is dissociated in solution in austenite leads to practically the same curve no definite decision can be made regarding the nature of the solution.

IV. The heat of transformation of austenite

When a homogeneous crystal of austenite containing 0.86% C is cooled below 721° C. it breaks up into Fe_3C and α-iron, liberating the heat of solution of α-iron and Fe_3C. The heat of solution of 1 g. of α-iron in austenite at 721° is approximately the same as the heat of transformation at 721°, which is 18.9 cal. according to Table I. The heat of solution of 1 g. of carbide was calculated in section III to be 30.2 cal, hence neglecting the small solubility of Fe_3C in α-iron and considering that 1 g. of austenite (0.86% C) gives 0.861 g. Fe and 0.139 g. Fe_3C the heat of transformation of 1 g. of pearlite at 721° is

$\Delta H = (0.139 \times 30.2 + 0.861 \times 18.9)$ cal $= 20.5$ cal. This agrees well with an unpublished datum of Esser (22.5 cal) but is somewhat higher than the value of 15.5 cal reported by Meuthen.[5]

V. The curves for the beginning and end of the separation of austenite from the melt, the so-called liquidus and solidus lines (lines BC and JE, Fig. 4).

The experimental observations on the position of these 2 lines are far from satisfactory and the thermal data which enter into the calculations involve large uncertainties. It must be emphasized, therefore, that in the discussion of this region of the diagram such quantitative deductions as are obtained are to be regarded as little more than intelligent guesses. The chief results derived from the application of thermodynamics to the problem are the relative probabilities of certain conditions assumed to exist on the basis of the data now available. When the serious handicaps imposed upon investigators working at these temperatures by the present limitations of technic have been

[5] Meuthen. *Ferrum,* Vol. 10, 1912/13, pages 1-21.

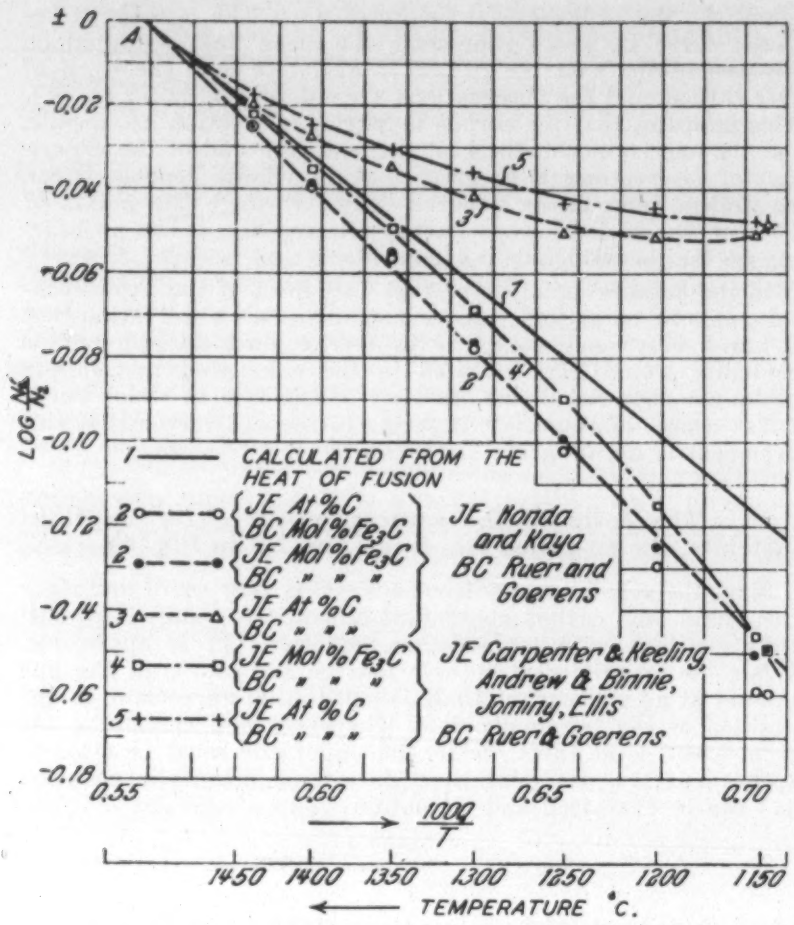

Fig. 3. Comparison of the experimental points on the BC and JE lines (Fig. 4) with the thermodynamic calculations.

overcome and reliable information is available, it may be possible to treat the problem with something like mathematical rigor by the use of the identical methods used in the analysis which follows. Until that time the conclusions which have been drawn can have little standing beyond that of interesting but tentative deductions.

In discussing these lines equations (5) and (6) are employed. In order to use the latter, it is necessary to justify the assumption that the heat of melting of γ-iron is independent of the temperature. A study of the existing total heat data shows that this approximation is substantially correct for while the actual results reported by various investigators show a wide variation it is usually found that for each set of data the slope of the heat content curve for the γ form is very nearly the same as that of the curve for liquid iron when extrapolated below the melting point. The heat of melting, which is the difference between the 2 curves is, therefore, practically constant with temperature. This is equivalent to saying that the specific heat of γ-iron is the same as that of molten iron which is not strictly true but the deviation is small and is well within the uncertainty in the heat of fusion.

The latent heat of melting of γ-iron is taken as equal to the sum of the heat of fusion of α-iron and the heat of the transformation. After studying the various values for these quantities reported in the literature, Körber and Oelsen estimate that the heat of fusion of γ-iron is 3760 cal./g.-atom. They also choose, more or less arbitrarily, the value of 1518° C. for the melting temperature of γ-iron. Using these data we get

$$\log \frac{N_2'}{N_2} = \frac{-3760}{2.303\,R}\left(\frac{1}{T} + \frac{1}{1518+273}\right) = \frac{-822.6}{T} - 0.4593 \quad (10)$$

where N'_2 is the mol fraction of free iron in the liquid saturated with C and N_2 is the mol fraction of free iron in austenite saturated with C.

The straight line corresponding to this equation is shown by the solid line in Fig. 3. The dotted curves are the lines calculated from the experimental data using 3 different assumptions as to the state of C in the system, namely that it is present as: (a) Fe_3C in the melt, Fe_3C in the austenite; (b) Fe_3C in the melt, C in the austenite; (c) C in the melt, C in the austenite. The fourth case of C in the melt, Fe_3C in the austenite has not been considered since the study shows that the condition of the C in the austenite has little effect on the calculations.

The lines 2 and 4 are based on assumptions (a) and (b) while 3 and 5 are based on (c). It is obvious from the 2 sets of points for line 2 that the condition of C in the austenite makes little difference in the location of the line, the important factor being the condition of C in the melt. It also appears that curves 2 and 4 come much closer to the straight line 1 than do 3 and 5 hence Körber and Oelsen concluded that the C in the melt is largely in the form of Fe_3C. They point out further that curves calculated from direct observation fall a little below curve 1, and they interpret this as indicating that the assumption that all the C is present as carbide gives an amount of Fe which is too low. They conclude, therefore, that the carbide in the melt is slightly dissociated, the degree of dissociation being represented by the distance between curve 2 or 4 and curve 1. Representing the dissociation constant by $K = \frac{[C]\,[Fe]^3}{[Fe_3C]}$ it is possible to draw further qualitative conclusions. (If the experimental data were better, quantitative results could also be secured by the same reasoning.) If curve 4 is accurate its curvature indicates that K decreases with increasing T corresponding to a positive heat of formation of

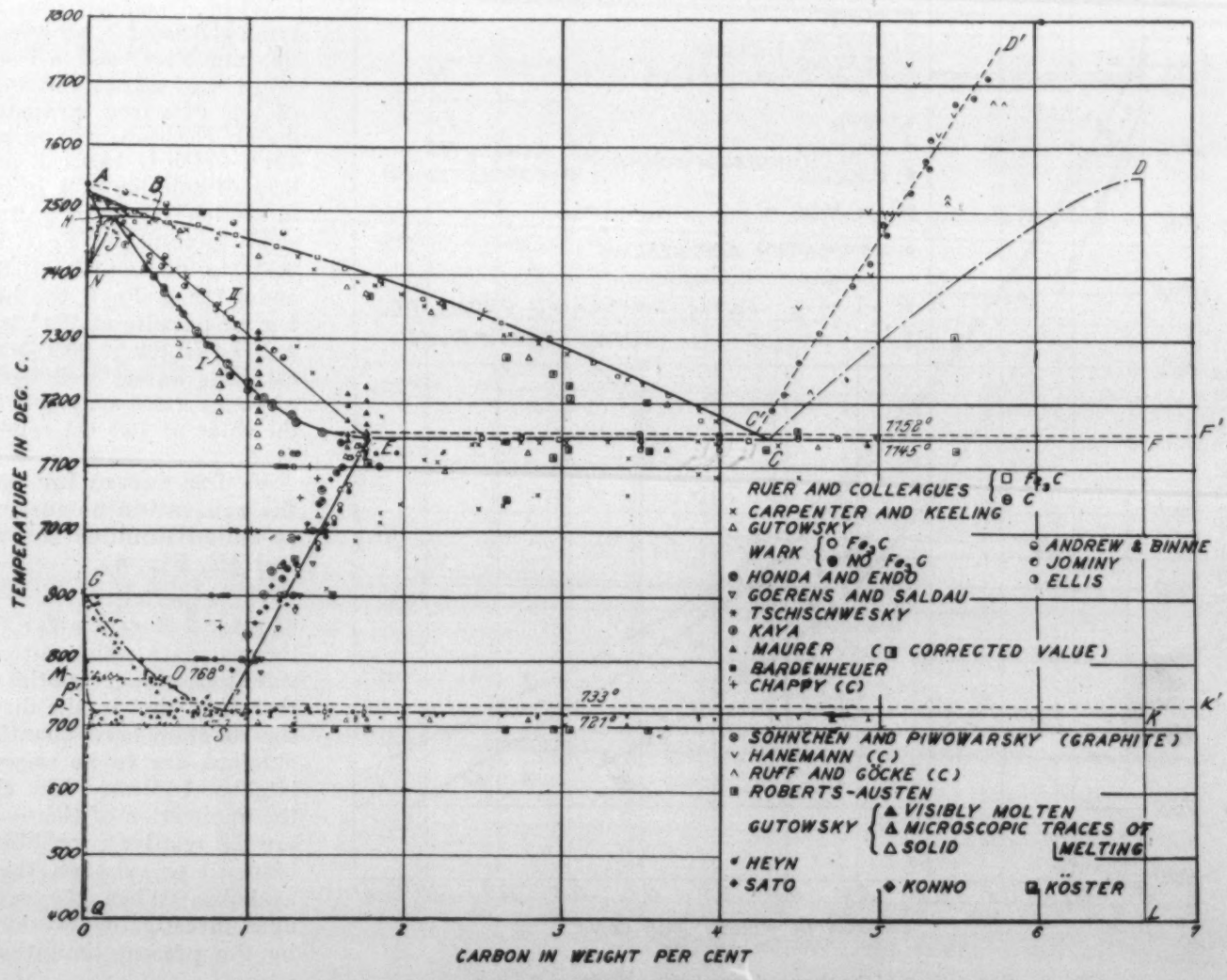

Fig. 4. Composite iron-carbon diagram showing the existing data

from 5000 to 7000 cal./mol ($3Fe + C = Fe_3C$ + 5000 to 7000 cal.). This is in fair agreement with the calorimetric results of Roth[6] and Watase[7]. The curvature of line 2 indicates, according to the authors, a negative heat of formation of approximately 10,000 cal./mol, a value which is not supported by experiment. While the data as they stand do lead to these deductions the uncertainty of the numerical values employed renders it unwise to lay too much stress upon them. No conclusive evidence is obtained from these curves for the state of the C dissolved in austenite.

It is also pointed out that curves 3 and 4 (Fig. 4) which are based on the line JE II seem to approach the straight line 1 asymptotically with decreasing C content whereas curves 2 and 5, based on the line JE I, always make an angle with 1. For this reason 3 and 4, and hence JE II, seem slightly more plausible than 2, 5 and JE I. The choice must be regarded as tentative, however.

Of the position of the lines AB, AH, HJB and of the location of the point B (Fig. 4) nothing definite can be said. This may be possible when the lines JE and BC are more definitely fixed and when the thermal data for pure iron are more accurately known but at present any detailed discussion of this part of the field is little more than speculation.

In the light of the preceding discussion it is evident that although the general features of the iron-carbon system are well established there are a number of details which require further study. It is the abstractor's belief, however, that the diagram of Körber and Oelsen (Fig. 4), which is consistent with all the existing evidence, with calculations based on thermal data as well as with direct observations, is now the most reliable interpretation of equilibrium conditions in this important system and that it can be substantially improved only as more accurate data become available.

SUMMARY

The several curves for the melting and transformation changes of iron-carbon alloys have been calculated from the heats of melting and transformation. The calculated curves are in general in fair agreement with direct observations (see Fig. 4). The A_3 line (GOS) is convex to the concentration axis and shows no break at 768° (A_2 point). The A_{cm} line (SE) is practically straight. A straight line (JE) for the upper limit of the austenite field appears more probable than a curved one. The heat of transformation of pearlite is calculated to be 20.5 cal./g. It is concluded that iron carbide has a positive heat of formation (heat is absorbed) of between 5,000 and 7,000 cal./mol. No conclusive evidence of the state of the carbon dissolved in austenite was obtained but there are indications that the carbon in the melt is chiefly present as Fe_3C which is very slightly dissociated.

[6] Roth. *Archiv für Eisenhüttenwesen*, Vol. 3, 1929/30, page 339.

[7] Watase. *Science Reports, Tôhoku Imperial University*, Vol. 17, 1928, page 1901.

Bibliography

The development of the iron-carbon diagram from its earliest stages to the present form can be conveniently traced in the following papers:

1888—Osmond. Transformation du fer et du carbone dans les fers, les aciers et les fontes blanches. Baudoin & Co., Paris, 1888.
1897-99—Roberts-Austen. 4th & 5th Report, Alloys Commission. *Proceedings Institution of Mechanical Engineers*, 1897, pages 31-100; 1899, pages 35-102.
1900—Roozeboom. *Zeitschrift für physikalische Chemie*, Vol. 34, 1900, pages 437-487.
1904—Heyn. *Zeitschrift für Elektrochemie*, Vol. 10, 1904, page 491.
1905—Charpy. *Comptes Rendus*, Vol. 141, 1905, page 948.
1906—Benedicks. *Métallurgie*, Vol. 3, 1906, pages 393, 425, 466.
1914—Ruer & Klesper. *Ferrum*, Vol. 11, 1913/14, page 257.
1917—Ruer & Goerens. *Ferrum*, Vol. 14, 1916/17, page 161.

In the following list under the various headings are grouped the most important papers, dealing with the location of the particular points or lines under discussion. The actual results are compared in Fig. 4, with the thermodynamically determined lines.

I. The equilibrium lines in the iron-carbon system.

A B curve for the initial separation of δ-ferrite from the melt
1. Carpenter & Keeling. *Journal Iron & Steel Institute*, Vol. 65, 1904, page 224.
2. Ruer & Klesper. *Ferrum*, Vol. 11, 1913/14, page 257.
3. Andrew & Binnie. *Journal Iron & Steel Institute*, Vol. 119, 1929, page 309.

A H curve for the end of the separation of δ-ferrite from the melt
Not determined. A and H are joined by an arbitrary line.

H N lines for the δ→γ transformation
1. Ruer & Klesper. *Ferrum*, Vol. 11, 1913/14, page 257.
2. Andrew & Binnie. *Journal Iron & Steel Institute*, Vol. 119, 1929, page 309.

J N not determined

B C curve for the beginning of separation of austenite from the melt
1. Roberts-Austen. 4th & 5th Reports, Alloys Commission. *Proceedings Institution of Mechanical Engineers*, 1897, pages 31-100; 1899, pages 35-102.

Table I.—Calculation of the line GOS from the heat content of pure iron as given by Oberhoffer and Grosse.

Temp. t°C.	Heat effect of the transformation at t°C. cal./g.	cal./g.-atom	$\frac{10^3}{2.303R}$	Values of $\int_T^{T_t} \frac{f(T)}{T^2} dT$	Iron content of the saturated solid solution atomic percent	Carbon content of the saturated austenite in wt. percent calcd. as mol. % Fe_3C	calcd. as atomic % C
906	5.8	325		—	100.00	—	—
900	5.9	330		0.31	99.93	0.015	0.015
880	6.0	336		1.39	99.68	0.068	0.069
860	6.4	358		2.56	99.41	0.124	0.127
840	6.7	375		3.83	99.12	0.184	0.191
820	7.1	397		5.22	98.80	0.250	0.259
800	7.6	426		6.75	98.46	0.320	0.335
790	8.1	454		7.59	98.27	0.360	0.376
780	9.2	515		8.54	98.05	0.399	0.424
775	10.3	576		9.08	97.93	0.425	0.451
770	11.7	655		9.70	97.79	0.452	0.483
760	14.8	829		11.21	97.45	0.517	0.558
755	16.0	896		12.10	97.25	0.557	0.603
750	17.2	963		13.06	97.04	0.596	0.651
740	18.2	1020		15.16	96.57	0.683	0.758
730	18.6	1042		17.38	96.08	0.775	0.866
720	18.9	1058		19.68	95.57	0.867	0.983
710	19.0	1064		22.06	95.05	0.957	1.102

2. Carpenter & Keeling. *Journal Iron & Steel Institute*, Vol. 65, 1904, page 224.
3. Gutowsky. *Métallurgie*, Vol. 6, 1909, page 737.
4. Ruer & Klesper. *Ferrum*, Vol. 11, 1913/14, page 257.
5. Ruer & Goerens. *Ferrum*, Vol. 14, 1916/17, page 161.
6. Andrew & Binnie. *Journal Iron & Steel Institute*, Vol. 119, 1929, page 309.

J E curve for the end of separation of austenite from the melt
1. Carpenter & Keeling. *Journal Iron & Steel Institute*, Vol. 65, 1904, page 224 (thermal analysis).
2. Gutowsky. *Métallurgie*, Vol. 6, 1909, page 737 (quenching—cooling experiments).
3. Kaya. *Science Reports Tôhoku Imperial University*, Vol. 14, 1925, page 529 (electrical resistance).
4. Ellis. *Journal Iron & Steel Institute, Carnegie Scholarship Memoirs*, Vol. 15, 1926, page 195 (malleability).
5. Honda & Endo. *Science Reports Tôhoku Imperial University*, Vol. 16, 1927, page 235 (magnetic susceptibility).
6. Andrew & Binnie. *Journal Iron & Steel Institute*, Vol. 119, 1929, page 309 (thermal analysis).
7. Jominy. *Transactions American Society for Steel Treating*, Vol. 16, 1929, page 372 (burning and overheating steel).

The values from 3 and 5 are in good agreement and the work of Gutowsky does not contradict them; the data of 1, 4, 6 and 7 lie at higher carbon content and show that J E is a straight line.

C D curve for the beginning of the separation of cementite from the melt
1. Roberts-Austen. 5th Report, Alloys Commission. *Proceedings Institution of Mechanical Engineers*, 1899, pages 35-102.
2. Schematically by Ruer. *Zeitschrift für anorganische und allgemeine Chemie*, Vol. 117, 1921, page 249.

C'D' curve for the initial separation of graphite from the melt
1. Hanemann. *Stahl und Eisen*, Vol. 31, 1911, page 333.
2. Ruff & Goecke. *Métallurgie*, Vol. 8, 1911, page 417.
3. Wittorf. *Zeitschrift für anorganische und allgemeine Chemie*, Vol. 79, 1911, page 1.
4. Ruer & Biren. *Zeitschrift für anorganische und allgemeine Chemie*, Vol. 113, 1920, page 98.

E S curve for the solubility of cementite in austenite
1. Roberts-Austen. 5th Report, Alloys Commission. *Proceedings Institution of Mechanical Engineers*, 1899, pages 35-102 (microscopic examination).
2. Carpenter & Keeling. *Journal Iron & Steel Institute*, Vol. 65, 1904, page 224 (thermal analysis).
3. Gutowsky. *Métallurgie*, Vol. 6, 1909, page 737 (chemical method).
4. Wark. *Métallurgie*, Vol. 8, 1911, page 704 (quenching and microscopic examination).
5. Goerens & Saldau. *Revue de la Société Russe de Métallurgie*, 1914, I, page 789; *Stahl und Eisen*, Vol. 38, 1918, page 15 (electrical resistance and hardness).
6. Tschischewsky & Schulgin. *Journal Iron & Steel Institute*, Vol. 95, 1917, page 189 (microscopic examination and chlorine etching).
7. Kaya. *Science Reports Tôhoku Imperial University*, Vol. 14, 1925, page 529 (electrical resistance).
8. Konno. *Science Reports Tôhoku Imperial University*, Vol. 12, 1923/24, page 125 (dilatometric studies).
9. Stäblein. *Stahl und Eisen*, Vol. 46, 1926, page 101 (dilatometric analysis).
10. Honda & Endo. *Science Reports Tôhoku Imperial University*, Vol. 16, 1927, page 235 (magnetic susceptibility).
11. Sato. *Technology Reports Tôhoku Imperial University*, Vol. 8, 1929, page 27 (thermal and dilatometric studies).

E' S' curve for the solubility of graphite in austenite
1. Mannesmann. *Verhandlungen der Vereinigung zur Beförderung des Gerwerbefleisses*, Vol. 58, 1879, page 31.
2. Royston. *Journal Iron & Steel Institute*, Vol. 51, 1897, page 166.
3. Charpy. *Comptes Rendus*, Vol. 145, 1907, page 1277.
4. Benedicks. *Métallurgie*, Vol. 5, 1908, page 41.
5. Howe. *Métallurgie*, Vol. 6, 1909, pages 65, 105. Gives summary of previous work.
6. Gutowsky. *Métallurgie*, Vol. 6, 1909, page 737.
7. Thomsen. *Dissertation, Technische Hochschule, Berlin*, 1910.
8. Ruer & Iljin. *Métallurgie*, Vol. 8, 1911, page 97.
9. Ruer & Goerens. *Ferrum*, Vol. 14, 1916/17, page 161 (1.32% C at 1120° C.).

10. Söhnchen & Piwowarsky. *Archiv für Eisenhüttenwesen,* Vol. 5, 1931/32, page 111.
Charpy, Benedicks, Thomsen, Ruer & Iljin and Ruer & Goerens found the solubility of graphite in austenite to be very much less than that of cementite, while Royston, Gutowsky and Söhnchen & Piwowarsky observed no difference. The position of the E′ S′ line is therefore uncertain.

G O S curve for the separation of ferrite from austenite
1. Roberts-Austen. 5th Report, Alloys Commission. *Proceedings Institution of Mechanical Engineers,* 1899, pages 35-102.
2. Carpenter & Keeling. *Journal Iron & Steel Institute,* Vol. 65, 1904, page 224.
3. Heyn. *Verhandlung der Vereinigung zur Beförderung des Gewerbefleisses,* 1904, page 355. See also Heyn. Die Theorie der Eisen-Kohlenstoff-Legierungen, Julius Springer, Berlin, 1924, page 15.
4. Goerens & Meyer. *Métallurgie,* Vol. 7, 1910, page 307.
5. Rümlein & Maire. *Ferrum,* Vol. 12, 1914/15, page 141.
6. Bardenheuer. *Ferrum,* Vol. 14, 1916/17, pages 129, 145.
7. Goerens & Saldau. *Stahl und Eisen,* Vol. 38, 1918, page 15.
8. Maurer & Hetzler. See Maurer, *Mitteilungen Kaiser-Wilhelm-Institut für Eisenforschung,* Vol. 1, 1920, page 38. Contains a critical examination of the literature.
9. Konno. *Science Reports, Tôhoku Imperial University,* Vol. 12, 1923/24, page 125.
10. Stäblein. *Stahl und Eisen,* Vol. 46, 1926, page 101.
11. Esser. *Stahl und Eisen,* Vol. 47, 1927, page 337.
12. Sato. *Technology Reports, Tôhoku Imperial University,* Vol. 8, 1929, page 27.

G P curve for the ending of the separation of ferrite from austenite
Not determined, only schematic.

P Q curve for the solubility of cementite in ferrite
1. Vamada. *Science Reports, Tôhoku Imperial University,* Vol. 15, 1926, page 851 (only schematic).
2. Köster. *Archiv für Eisenhüttenwesen,* Vol. 2, 1928/29, page 503.

II. The temperatures of three phase equilibrium

H J B equilibrium between melt high-temperature ferrite and austenite
1. Ruer & Klesper. *Ferrum,* Vol. 11, 1913/14, page 257 (1487°).
2. Andrew & Binnie. *Journal Iron & Steel Institute,* Vol. 119, 1929, page 309 (1495°).

E C F cementite eutectic
1. Roberts-Austen. 4th & 5th Reports, Alloys Commission. *Proceedings Institution of Mechanical Engineers,* 1897, page 31-100; 1899, pages 35-102 (1120° to 1140°).
2. Carpenter & Keeling. *Journal Iron & Steel Institute,* Vol. 65, 1904, page 224 (1110° to 1146°).
3. Gutowsky. *Métallurgie,* Vol. 6, 1909, page 737 (1093° to 1134°).
4. Ruer & Goerens. *Ferrum,* Vol. 14, 1916/17, page 161 (1145°).

E′ C′F′ graphite eutectic
1. Thomsen. *Dissertation, Technische Hochschule, Berlin,* 1910 (E′ S′ 1180°).
2. Ruer & Goerens. *Ferrum,* Vol. 14, 1916/17, page 161 (1152°).

P S K pearlite eutectoid
1. Roberts-Austen. 5th Report, Alloys Commission. *Proceedings Institution of Mechanical Engineers,* 1899, pages 35-102 (700°).
2. Carpenter & Keeling. *Journal Iron & Steel Institute,* Vol. 65, 1904, page 224 (688° to 717°).
3. Heyn. *Zeitschrift für Elektrochemie,* Vol. 10, 1904, page 491 (700° to 705°).
4. Meuthen. *Ferrum,* Vol. 10, 1912/13, page 1 (710°).
5. Rümlein & Maire. *Ferrum,* Vol. 12, 1914/15, page 141 (718°).
6. Bardenheuer. *Ferrum,* Vol. 14, 1916/17, pages 129, 145 (710° to 721.5°).
7. Ruer & Goerens. *Ferrum,* Vol. 14, 1916/17, page 161 (721 ± 3°).
8. Saldau. *Stahl und Eisen,* Vol. 38, 1918, page 39 (707°).
9. Goerens & Saldau. *Stahl und Eisen,* Vol. 38, 1918, page 15 (700° to 710°).
10. Maurer. *Mitteilungen Kaiser-Wilhelm-Institut für Eisenforschung,* Vol. 1, 1920, page 38 (698° to 716°).
11. Konno. *Science Reports, Tôhoku Imperial University,* Vol. 12, 1923/24, page 125 (721 ± 3°).
12. Esser. *Stahl und Eisen,* Vol. 47, 1927, page 337 (708° ↑ 736°↑ mean 722°).
13. Sato. *Technology Reports, Tôhoku Imperial University,* Vol. 8, 1929, page 27 (726° ↕).

* ↓cooling ↑heating

P′ S′ K′ graphite eutectoid
1. Ruer. *Zeitschrift für anorganische und allgemeine Chemie,* Vol. 117, 1921, page 249 (733°).
Rapatz & Pollach. *Stahl und Eisen,* Vol. 44, 1924, page 1509.
2. Hayes, Flanders & Moore. *Transactions American Society for Steel Treating,* Vol. 5, 1924, page 183 (758°).

III. The concentrations of the principal points.

H Saturation concentration of C in high-temperature ferrite
Ruer & Klesper. *Ferrum,* Vol. 11, 1913/14, page 257 (0.07% C).

J Saturation concentration of C in austenite
Ruer & Klesper. *Ferrum,* Vol. 11, 1913/14, page 257 (0.18% C approx.)

B Saturation concentration of ferrite and austenite in the melt
1. Ruer & Klesper. *Ferrum,* Vol. 11, 1913/14 page 257 (0.36% C).
2. Andrew & Binnie. *Journal Iron & Steel Institute,* Vol. 119, 1929, page 309 (0.71% C).
3. Ruer & Goerens. *Ferrum,* Vol. 14, 1916/17, page 161 (0.6% C. Extrapolated from BC).

E Greatest possible saturation concentration of Fe_3C in austenite
Intersection of SE, ECF and JE (1.7% C).

E′ Greatest possible saturation concentration of graphite in austenite
Intersection of E′S′, JE, E′C′F′. Very doubtful on account of the uncertainty in E′S′.
1. Gutowsky. *Métallurgie,* Vol. 6, 1909, page 737 (1.713% C at 1135°; 1.704% C at 1120°).
2. Ruer & Iljin. *Métallurgie,* Vol. 8, 1911, page 97 (1.25% C at 1120°).
3. Ruer & Goerens. *Ferrum,* Vol. 14, 1916/17, page 161 (1.32% C at 1120°).
4. Söhnchen & Piwowarsky. *Archiv für Eisenhüttenwesen,* Vol. 5, 1931/32, page 111 (1.84% C at 1100°. Extrapolated from E′S′ 1.75% C at 1100°).

C Eutectic concentration. Intersection of E C F and B C
1. Benedicks. *Métallurgie,* Vol. 3, 1906, page 466 (4.2% C approx.).
2. Gutowsky. *Métallurgie,* Vol. 6, 1909, page 737 (about 4.2% C).
3. Ruer & Goerens. *Ferrum,* Vol. 14, 1916/17, page 161 (4.2% C).
4. Ruer & Biren. *Zeitschrift für anorganische und allgemeine Chemie,* Vol. 113, 1920, page 98 (4.3%C).

C′ Concentration of the graphite eutectic. Intersection of C′D′ and E′C′F′
1. Roberts-Austen. 5th Report, Alloys Commission. *Proceedings Institution of Mechanical Engineers,* 1899, pages 35-102 (4.3% C).
2. Carpenter & Keeling. *Journal Iron & Steel Institute,* Vol. 65, 1904, page 224 (4.3% C).
3. Ruer & Goerens. *Ferrum,* Vol. 14, 1916/17, page 161 (E′C′F′ 4.15% C).
4. Ruer & Biren. *Zeitschrift für anorganische und allgemeine Chemie,* Vol. 113, 1920, page 98 (C′D′ 4.25% C).

S Eutectoid concentration
1. Roberts-Austen. 5th Report, Alloys Commission. *Proceedings Institution of Mechanical Engineers,* 1899, pages 35-102 (0.85% C).
2. Carpenter & Keeling. *Journal Iron & Steel Institute,* Vol. 65, 1904, page 224 (0.89% C).
3. Heyn. *Zeitschrift für Elektrochemie,* Vol. 10, 1904, page 491 (0.93% C).
4. Goerens & Meyer. *Métallurgie,* Vol. 7, 1910, page 307 (0.95% C).
5. Rümlein & Maire. *Ferrum,* Vol. 12, 1914/15, page 141 (0.75% C).
6. Meuthen. *Ferrum,* Vol. 10, 1912/13, page 1 (0.9% C).
7. Bardenheuer. *Ferrum,* Vol. 14, 1916/17, pages 129, 145 (0.86% C).
8. Maurer. *Mitteilungen des Kaiser-Wilhelm-Institut für Eisenforschung,* Vol. 1, 1920, page 38 (0.75% C).
9. Konno. *Science Reports, Tôhoku Imperial University,* Vol. 12, 1923/24, page 125 (0.91% C).
10. Stäblein. *Stahl und Eisen,* Vol. 46, 1926, page 101 (0.86%C).
11. Esser. *Stahl und Eisen,* Vol. 47, 1927, page 337 (0.86% C).
12. Sato. *Technology Reports, Tôhoku Imperial University,* Vol. 8, 1929, page 27 (0.86% C).

S′ Concentration of the graphite eutectoid. Intersection of E′S′ and P′S′K′, and is as uncertain as both these lines.
1. Ruer. *Zeitschrift für anorganische und allgemeine Chemie,* Vol. 117, 1921, page 249. (0.7% C at 733°).
2. Söhnchen & Piwowarsky. *Archiv für Eisenhüttenwesen,* Vol. 5, 1931/32, page 111 (0.85% C at 733°). (The temperature is taken from Ruer.)
3. Royston. *Journal Iron & Steel Institute,* Vol. 51, 1897, page 166 (0.85 to 0.90% C at 720°).
4. Thomsen. *Dissertation, Technische Hochschule, Berlin,* 1910 (0.9% C at 750° after 550 hrs.).

P Maximum solubility of Fe_3C in ferrite
1. Scott. *Journal American Institute Electrical Engineers,* Vol. 43, 1924, page 1066 (0.04% C).
2. Hanemann. *Zeitschrift Verein deutscher Ingenieure,* Vol. 71, 1927, pages 246 (0.03% C).
3. Köster. *Archiv für Eisenhüttenwesen,* Vol. 2, 1928/29, page 503 (0.043% C at 710°).

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(Editorial—Continued from page A9)

With sufficient financial and legal backing, most any patent might be made into a big stick. If the validity of a patent is to be primarily judged by the size of the stick that is shaken, where are logic and justice?

Of course, there is a possibility that some time in the future, sensible judges will figure that, after all, the big stick is not proof, that they will decide cases themselves instead of letting the fact that some lawyers have advised their clients to pay rather than to fight, decide it for them, and so go back to the former, more reasonable precedent. But can we wait for that? Industry cannot talk to judges, but it can talk to law makers. The danger signal should be heeded and steps taken to define unmistakably what is and is not proof of validity.

In the absence of definite law, the latest court decision tends to be taken as the equivalent of law, but an evil trend can be corrected by the enactment of definite law, and the people have a right to ask their representatives for corrective legislation.

There is law, there is gospel, and there are apocryphal decisions. It is time to revert to the first and avoid the last.—H. W. Gillett

Effect of Nickel Upon Stability of Iron Carbide and Upon Microscopic Structure of White Cast Iron Compositions*

By MILO J. STUTZMAN**

Introduction

The investigation reported in this paper was undertaken with the object of determining the quantitative effect of nickel upon the rate of graphitization of iron carbide and upon the microscopic structure of white cast iron of malleable iron compositions.

A study of the literature revealed much information on the effects and use of nickel in gray cast iron and steel, but only a limited amount of information on the effects of nickel in malleable iron. Some of the data given on the effect of nickel on the rate of graphitization are confusing if not misleading.

Dangerfield, Johnson and Taylor[1] studied the effect of, from 0 to 3.66% of nickel on the physical properties of cupola melted white heart malleable iron with the special reference to thin walled material. Twenty test bars (8 × 1 × 3/16 in.) were cast from each composition studied. These were annealed in the commercial 5½ day anneal, packed in hematite ore. With the addition of 2.53% nickel the tensile strength was increased from 41000 lbs./in.² to 51000 lbs./in.² Further increase in nickel content decreased the tensile strength to 14300 lbs./in.² for 3.66% nickel. A nickel content of 0.70% gave the best values for a bend test around a 1 inch radius.

Wentrup and Stenger[2] made additions of 0.53, 0.98 and 1.78% Ni to a base metal containing 3.00% C; 0.20% Mn; 0.02% S; and 0.06% P. They studied the graphitization of these alloys by use of a dilatometer. Less than 1.00% Ni had no marked effect on the rate of graphitization, but the 1.78% Ni alloy showed a marked increase in the rate at which the iron carbide was decomposed. Photomicrographs were included to show the effect of nickel in decreasing the size of the graphite nodules.

Thiery[3] made additions of 0.5, 1.40 and 2.50% Ni to an iron containing 3.05% C and 0.40% Si. By means of a dilatometer he found that Ni lowered the temperature of the formation of graphite and also the temperature of the A_1 transition. Murry[4] studied the range of 0 to 3.53% Ni in both crucible and cupola melted malleable iron. The anneal used, gave a black heart malleable iron with a ferrite shell and between these areas, there was an area composed of ferrite and pearlite. In the alloy containing 1.00% Ni the completely decarburized area was eliminated, but was present in alloys of higher contents. The tensile strength was increased by Ni additions to a maximum with 2.50% Ni. The elongation was reduced.

Schwartz[5] determined the amount of graphite formed at stated times of anneal at constant temperatures, for commercial malleable irons containing Ni additions in the range of 0 to 2.50% Ni. In all but one of the alloys reported Ni increased the amount of graphite formed, during the stated time of anneal at a constant temperature. From these data Schwartz proceeded to show by mathematical deductions that the effects of Ni and Si on the rate of breakdown of free iron carbide are not additive. In this paper it will be shown that the rate of graphitization in the first stage bears a definite relation to the Ni content of the alloy.

Experimental Methods and Procedure

The alloys reported in this paper were prepared by melting weighed quantities of white cast iron and electrolytic Ni in an Ajax induction electric furnace and casting into ⅝ in. test bars in dry sand molds. The white cast iron used was of the following composition: 2.25 to 2.35% C; 0.90 to 0.96 Si; 0.033

*This paper is based upon the content of a thesis submitted to the Graduate Faculty of Iowa State College, Ames, Iowa, by the author in partial fulfillment of the requirements for the degree of Doctor of Philosophy.

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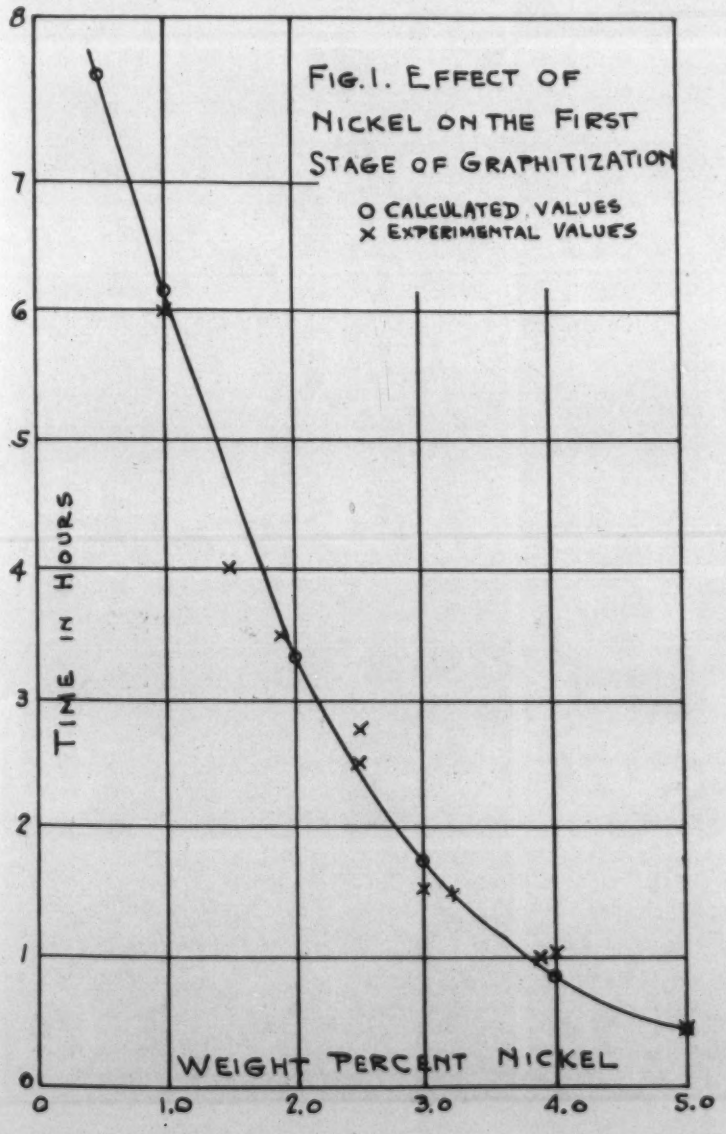

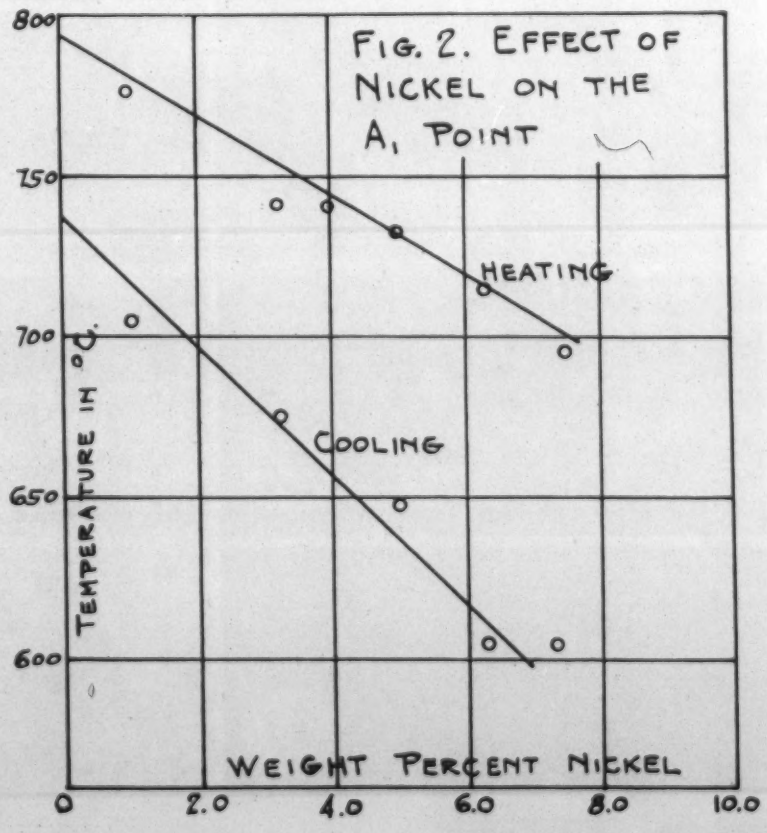

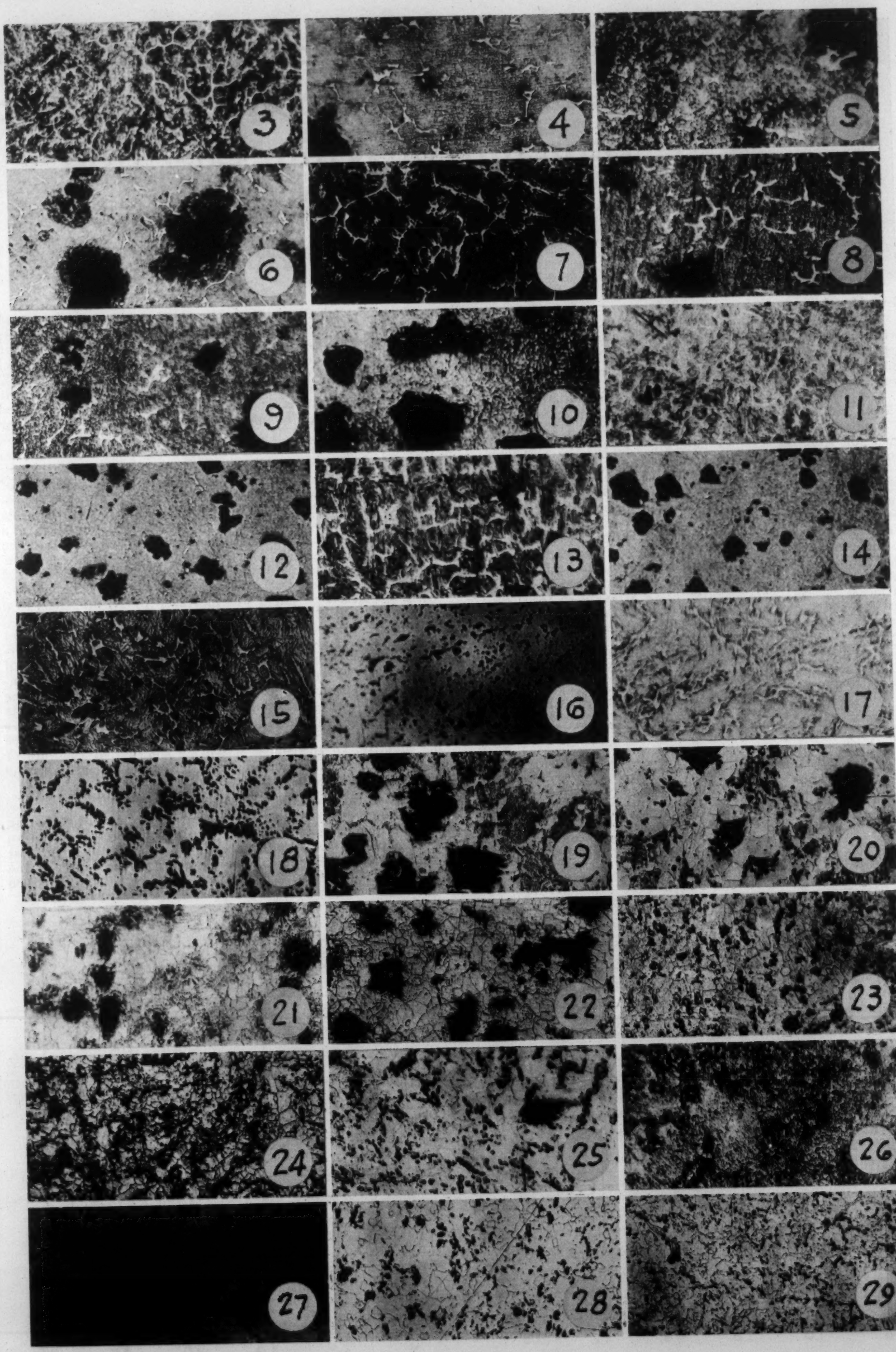
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Fig. 3 Sample 16 with 0.0 % Ni as cast. Nitric acid etched. 100X.

Fig. 4 Sample 16. 0.0 % Ni. Annealed for 3 hrs. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 5 Sample 16. 0.0 % Ni. Annealed for 4 hrs. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 6 Sample 16. 0.0 % Ni. Annealed for 6 hrs. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 7 Sample 17. 1.9 % Ni. As cast. Nitric acid etched. 100X.

Fig. 8 Sample 17. 1.9 % Ni. Annealed for 2 hrs. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 9 Sample 17. 1.9 % Ni. Annealed for 3 hrs. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 10 Sample 17. 1.9 % Ni. Annealed for 4 hrs. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 11 Sample 18. 2.5 % Ni. As cast. Nitric acid etched. 100X.

Fig. 12 Sample 18. 2.5 % Ni. Annealed for 2 hrs. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 13 Sample 19. 3.2 % Ni. As cast. Nitric acid etched. 100X.

Fig. 14 Sample 19. 3.2 % Ni. Annealed for 2 hrs. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 15 Sample 20. 3.9 % Ni. As cast. Nitric acid etched. 100X.

Fig. 16 Sample 20. 3.9 % Ni. Annealed for 1 hr. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 17 Sample 10. 5.9 % Ni. As cast. Nitric acid etched. 100X.

Fig. 18 Sample 10. 5.9 % Ni. Annealed for ½ hr. at 936°C. Air cooled. Nitric acid etched. 100X.

Fig. 19 Sample 16. 0.0 % Ni. Annealed at High Temperature for 16 hrs. Then cooled to 648°C. (1200°F.) in 10 hrs., and in 6 hrs. more to 538°C. (1000° F.) Then air cooled. Nitric acid etched. 100X.

Fig. 20 Sample 17. 1.9 % Ni. Same heat treatment as Fig. 19. Nitric acid etched. 100X.

Fig. 21 Sample 18. 2.5 % Ni. Same heat treatment as in Fig. 19. Nitric acid etched. 100X.

Fig. 22 Sample 19. 3.2 % Ni. Same heat treatment as in Fig. 19. Nitric acid etched. 100X.

Fig. 23 Sample 20. 3.9 % Ni. Same hea treatment as in Fig. 19. Nitric acid etched. 100X.

Fig. 24 Sample 10. 5.9 % Ni. Same heat treatment as in Fig. 19. Nitric acid etched. 100X.

Fig. 25 Sample 12. 7.3 % Ni. Same heat treatment as in Fig. 19. Nitric acid etched. 100X.

Fig. 26 Sample 14. 8.8 % Ni. Same heat treatment as in Fig. 19. Nitric acid etched. 100X.

Fig. 27 Sample 31. 3.46 % Ni. Annealed for 2 hrs. at high annealing temperature and cooled to 593°C. (1100°F.) in 6 hrs. Then air cooled. Nitric acid etched. 100X.

Fig. 28 Sample 25. 4.18 % Ni. Annealed for 1 hr. at high annealing temperature and cooled to 593°C. (1100°F.) in 6 hrs. Then air cooled. Nitric acid etched. 100X.

Fig. 29 Sample 9. 4.0 % Ni. Annealed for 1 hr. at high annealing temperature and cooled to 593°C. (1100°F.) in 6 hrs. Then air cooled. Nitric acid etched. 100X.

S; 0.21 Mn; 0.155 P. The melts were made in magnesia crucibles, stirred twice with an Armco iron rod and poured at 1500°C. (2732°F.) The bars were allowed to cool to a temperature below the A_1 transition in the molds. Chemical analysis of the alloys showed the Ni and C contents to be within experimental error of the calculated values.

The method used for determining the rate of graphitization during the first stage, was to determine the time required for the disappearance of free iron carbide. To do this a large number of pieces of each composition under study were placed in a Hump annealing furnace, and at half hour intervals, one of each was removed from the furnace, cooled in air, polished and examined microscopically for free iron carbide. No attempt was made to determine the time closer than the half hour intervals. The annealing temperature used in this work was 936°C. (1717°F.) Micrographs (Figs. 3 to 6) show the progress of graphitization for the base white iron.

The method used for following the rates of graphitization in the second stage was to cool the alloys from the temperature of the first stage anneal to a temperature just below the eutectoid at a rate sufficiently slow so that free iron carbide would not separate, and then to determine the time required for complete graphitization. The effect of nickel on the A_1 transition for these alloys was determined for 0 to 7% Ni. Heating and cooling curves were run, using a L & N differential critical point apparatus. Alloys containing more than 2.5% Ni graphitized so rapidly in the critical range that it was not necessary to retard the cooling at the eutectoid temperature. The rates of graphitization were then studied by varying the rate of cooling through the critical range and determining the combined C microscopically.

From the data thus obtained on the rates of graphitization in both stages, the time required for complete graphitization to ferrite and graphite was computed. Alloys were then annealed in accordance with these estimations and examined under the microscope.

Presentation of Results

Data are shown in Table I giving the time required for the completion of the first stage of graphitization.

Table I. Time for Completion of the First Stage of Graphitization for Varied Percents of Nickel, at 936°C. (1717°F.)

Sample Number	Weight % Nickel	Time of First Stage
16	0.00%	11-12 hrs.
15	1.00	6
8	1.50	4
17	1.90	3.5
2	2.50	3
18	2.50	2.5
3	3.00	1.5
19	3.20	1.5
20	3.90	1.0
9	4.00	1.0
6	4.00	1.0
5	5.00	0.5
10	5.90	0.5
11	6.30	0.5
12	7.30	0.5
14	8.80	0.5
13	10.30	0.5

No attempt was made to determine the time to less than half hour intervals. Those samples marked 0.5 hours contained no free carbide after the first half hour at the annealing temperature.

Kikuta[6] showed that the effect of Si on the time required for the first stage of graphitization could be expressed by the equation: $t = A \cdot S^B$, where t is time in hours, A and B constants and S the weight percent of silicon.

It was found that the effect of Ni on the time required for the first stage of graphitization may be expressed by the equation:

$$t = A \cdot e^{BN} \qquad (1)$$

Where t is time in hours, A and B are constants, N the weight percent of Ni, and e the natural logarithm.

The data for samples 15, 17, 18, 19, and 20 representing a range of composition of from 1.00 to 3.90% Ni were used in the evaluation of the constants A and B in Equation 1. Substituting, in Equation 1, the average values of A and B thus obtained, Equation 1 becomes,

$$t = 12.0 \cdot e^{-0.654N} \qquad (2)$$

Values of N were then substituted in Equation 2 and solved for t. The values thus obtained for t are shown in Table II.

Table II. Calculated Time for the First Stage of Graphitization.

N	t	N	t
0.0	12.	3.0	1.74
0.5	7.86	4.0	0.86
1.0	6.18	5.0	0.45
2.0	3.34	6.0	0.24

In Fig. 1 the calculated values of t are plotted on the curve, the calculated values being marked ○. The experimental values are plotted on the same figure and marked ×. It is thus shown that the experimental data agree very closely with those calculated from the equation. However this equation would not give concordant values for alloys that were far variant in silicon or carbon contents from those used in this work.

Kikuta[6] studied the effect of Si on the time required for the first stage of graphitization. An alloy containing 2.00% C and 0.98% Si required 11 hours and an alloy containing 2.00% C and 2.20% Si, representing an increase of 1.22% Si, required 1 hour to complete the first stage of graphitization. In this paper it is shown that, for an alloy containing 2.30% C and 0.94% Si, 3.90 to 4.00% Ni is required to reduce the time of the first stage of graphitization from 12 hours to 1 hour. Thus approximately 3 times as much Ni as Si is required for an equivalent reduction in the time required for the graphitization of the free iron carbide.

Micrographs (Figs. 3 to 18) show the micro structures of alloys containing 0 to 6.0% Ni, before and after annealing, and also the decrease in time required for the first stage of graphitization with increasing Ni content. The alloys were also examined under the microscope in the as cast and unetched condition, thus showing the absence of primary graphite. The micrographs show also the reduction in the size of the graphite nodules to be in proportion to the amount of Ni present up to about 3.0% Ni. At about 3.5% Ni there occurs an abrupt decrease in the size of the graphite particles.

The effect of Ni on the A_{r1} and the A_{c1} point for the Fe-C-Ni alloys containing 0.95% Si is shown in Fig. 2. The values

are about 54°C. (98°F.) higher than those obtained by Kase[7] for pure Fe-C-Ni alloys. Hayes and Diederichs[8] found that 1.0% Si raised the A_1 point approximately 59°C. (108°F.). This would indicate that the effect of Ni on the temperature of the A_1 point to be of the same magnitude in 1% Si alloys as in pure Fe-C alloys.

The effect of Ni on the rate of graphitization in the critical range is shown by the micrographs (Fig. 19 to 26). The alloys were heated at the high temperature for sufficient time to complete the first stage of graphitization, and then cooled to 648°C. (1200°F.) in 10 hours. The cooling rate was then retarded so that 5 hours more were required to cool to 538°C. (1000°F.), at which time they were removed from the furnace and cooled in air. Sample 19 (Fig. 22) containing 3.2% Ni, showed only small areas of pearlite as compared to sample 20 (Fig. 23) containing 3.9% Ni which sample was completely graphitized. Sample 16 (Fig. 19) contained no Ni and showed large areas of pearlite. Sample 14 (Fig. 26) contained 8.80% Ni and showed graphite in a martensitic matrix.

The micrographs included in this paper show also that Ni has little or no effect on the microstructure of the free iron carbide in the white cast iron. As might be anticipated from the effect of Ni in steel and its solubility in the γ-iron solid solution, the matrix of white cast iron may be changed to martensite or to austenite by the addition of Ni. The micrographs show also that in the range of 0 to 3.0% Ni refines the grains of the ferrite and increases the number of C nuclei in the malleabilized cast iron. At about 3.5% Ni, there is a definite change in the microstructure of the malleable iron product. The number of C nuclei are greatly increased and are not evenly distributed throughout the material. This change occurs simultaneous with the change to martensitic matrix in the white iron, and would indicate a greater number of C nuclei in the martensitic material. Thus the C is precipitated at a greater number of centers.

Grateful acknowledgment is made to Dr. Anson Hayes, Director of Research, The American Rolling Mill Co., Middletown, Ohio. The work which is reported in this paper was done in the Division of Physical Chemistry, at Iowa State College in the general field of factors which affect rates of graphitization of white cast iron, under the direction of Dr. Hayes.

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Beg Pardon

In the Handy & Harman ad on page 3 of our March issue the omission of a paragraph, and lack of proper spacing, after Item No. 6, made the text read as if the "Sil-Fos" Brazing Alloy patent was based upon its corrosion-resistance. Handy & Harman inform us that the patent covers the unique composition and brazing qualities of the alloy, rather than any one of its individual characteristics.

LABORATORY EVENING

Quiet: chill fog and thin occasional rain.
The bottles on the Chem Lab shelves
Reflect the one light of the watchman's round.
Below, the microscopic bench is shrouded.
The work shop, littered with odd jobs,
Salvage of orphan hopes shop orders dead;
Half grown ideas, abandoned suddenly
As were the household tasks in Pompeii
When dread Vesuvius lashed out in anger
Of lava flow and boiling mud and ash.
Below again, the fatigue machines are still:
The Haigh, the R. R. Moore, the wire, and rope
No longer stretch their S. N. curves
Slowly and slower yet to logarithmic time.
In their sealed blackness the photographic films
Eagerly waiting the flash of impregnation.
A hundredth second ecstasy, and then
The slow baths and gradual development
And finally another flash, a twentieth second long
To recreate upon a distant screen the supreme moment
Of a past existence. In waiting idleness
Volt and ammeter, resistance, inductance capacity
Oscillographs and galvanometers, ready to wire up
In some new net to snare
A minute change in the magnetic state
Which indicates the subtle temper
Of impact tested steel. And all these instruments
Quiet, but yet alive, waiting but a touch
To spring to life, and quick, responsive
Infinitely flexible, report the worth
Of some new hope.
On some such night, with cheese and Swedish bread
And milk, and maybe C. P. alcohol
Will come the combination, the idea
So simple as to seem self-evident, and therefore
Deepest hidden, to which all tests
Will answer, "this is right" and steadily
With cheerful purr or rhythmic swing and click
The tools of research can move on once more.

ALFRED V. DE FOREST
Feb. 1933

C. B. Boyne will succeed C. B. Templeton as Assistant to the President of the Ludlum Steel Company. As recently announced, Mr. Templeton is now Assistant to the Vice President in Charge of Sales.

Mr. Boyne first joined the Ludlum organization in 1913 and has ever since been identified with the marketing of Ludlum products. He will now be located at the executive offices of the company at Watervliet, New York.

Raymond L. Collier, for a number of years Assistant Managing Director of the Steel Founders' Society, was recently appointed Managing Director of that organization to succeed Granville P. Rogers, who severed his connection with the Society to become affiliated with the Paper Cup Manufacturers Institute, New York City.

O. B. J. Fraser, formerly in charge of the International Nickel Company's Research Laboratory at Bayonne, N. J., is now located in the Development and Research Department at the company's main offices in New York, where he will be in charge of developments in the uses of nickel and nickel alloys in the oil, gas and coke industries, and in the industrial uses of non-metallic compounds of nickel.

Norman B. Pilling, formerly in charge of metallurgical research, has been placed in charge of the Research Laboratory. **E. M. Wise**, whose work heretofore has been largely associated with precious metals, has been appointed assistant.

Dr. Vsevolod N. Krivobok, professor of metallurgy at the Carnegie Institute of Technology, Pittsburgh, Pa., has accepted the invitation of the board of directors of the American Society for Steel Treating to deliver the Campbell Memorial lecture during the annual convention of the society in 1934.

Dr. Krivobok studied at the Polytechnical Institute at St. Petersburg. Following the Russian revolution he took up the study of metallurgy at Harvard where he received the B.S., M.S., and Sc.D. degrees.

Last year the Campbell lecture was delivered by **E. C. Bain** of the research laboratory of the U. S. Steel Corporation. This year **H. J. French** of the International Nickel Company and a member of METALS & ALLOYS Editorial Advisory Board will be the speaker.

Erich Fetz, formerly with the Scalex Products Co., in Brooklyn, has joined the staff of the Research Laboratory of the International Nickel Co., Bayonne, N. J.

Fred J. Esslinger has joined the technical staff of the Champion Rivet Company of Cleveland. He will be engaged in various research capacities, devoting his efforts primarily to the development of welding rod coatings for particular applications. Mr. Esslinger received his technical training at Friedrich Wilhelm and the Technical Universities of Berlin. He was formerly with the Petroleum Iron Works of Sharon, Pa., in charge of its metallurgical laboratory and welding rod developments.